
Supplementary information

Strong yet ductile nanolamellar high-entropy alloys by additive manufacturing

In the format provided by the
authors and unedited

Supplementary Information

Strong yet ductile nanolamellar high-entropy alloys by additive manufacturing

Jie Ren^{1,9}, Yin Zhang^{2,9}, Dexin Zhao³, Yan Chen⁴, Shuai Guan¹, Yanfang Liu¹, Liang Liu¹, Siyuan Peng¹, Fanyue Kong¹, Jonathan D. Poplawsky⁵, Guanhui Gao⁶, Thomas Voisin⁷, Ke An⁴, Y. Morris Wang⁸, Kelvin Y. Xie³, Ting Zhu^{*2}, Wen Chen^{*1}

¹Department of Mechanical and Industrial Engineering, University of Massachusetts, Amherst, MA 01003, USA

²George W. Woodruff School of Mechanical Engineering, Georgia Institute of Technology, Atlanta, GA 30332, USA

³Department of Materials Science and Engineering, Texas A&M University, College Station, TX 77843, USA

⁴Neutron Scattering Division, Oak Ridge National Laboratory, PO Box 2008 MS6475, Oak Ridge, TN 37831, USA

⁵Center for Nanophase Materials Sciences, Oak Ridge National Laboratory, Oak Ridge, TN 37831, USA

⁶Rice Electron Microscopy Centre, Department of Materials Science and Nano-Engineering, Rice University, Houston, TX 77005, USA

⁷Materials Science Division, Lawrence Livermore National Laboratory, Livermore, CA 94550, USA

⁸Department of Materials Science and Engineering, University of California, Los Angeles, CA 90095, USA

⁹These authors contributed equally: Jie Ren, Yin Zhang.

*Corresponding authors: T.Z. (ting.zhu@me.gatech.edu); W.C. (wenchen@umass.edu)

Table of Contents

S1 OPTIMIZATION OF PROCESSING PARAMETERS.....	4
Supplementary Fig. 1 Normalized processing diagram for L-PBF of AlCoCrFeNi _{2.1}	5
Supplementary Table 1 L-PBF processing parameters for optimized EHEA samples.	6
Supplementary Fig. 2 Secondary electron images of the cross-section views of the optimized samples A and B.	6
S2 MICROSTRUCTURAL CHARACTERIZAION	7
Supplementary Fig. 3 Nearly random orientations in shape and crystallography of nanolamellar eutectic colonies in AM AlCoCrFeNi _{2.1} EHEA.	7
Supplementary Fig. 4 Microstructural characterization of the AlCoCrFeNi _{2.1} powder feedstock.	8
Supplementary Table 2 Chemical compositions of AlCoCrFeNi _{2.1} powders and as-printed sample measured by ICP-MS and IGA.	8
Supplementary Fig. 5 Microstructural characterization of the chemically-modulated nanostructures within BCC lamellae of the as-printed AlCoCrFeNi _{2.1} EHEA.	9
Supplementary Fig. 6 Fracture morphologies of the as-printed AlCoCrFeNi _{2.1} EHEA.	10
Supplementary Fig. 7 Meso-scale deformation structures of the as-printed AlCoCrFeNi _{2.1} EHEA imaged by PED.	11
S3 <i>IN SITU</i> NEUTRON DIFFRACTION MEASUREMENTS UNDER TENSION	12
1. Experimental set-up.....	12
Supplementary Fig. 8 Schematic diagram of <i>in situ</i> neutron diffraction tensile testing set-up.....	12
2. Determination of reference $d_{0,hkl}$ and residual stress	12
Supplementary Fig. 9 Determination of a “stress-free” AM AlCoCrFeNi _{2.1} EHEA sample through thermal annealing	14
Supplementary Fig. 10 Characterization of microscale residual stresses through <i>in situ</i> neutron diffraction measurement of lattice strains in AM AlCoCrFeNi _{2.1} EHEA.	15
3. Lattice strain evolution along transverse direction.....	16
Supplementary Fig. 11 Evolution of the orientation-specific elastic lattice strain over macroscopic true stress for FCC- $\{111\}$, $\{200\}$, $\{220\}$, $\{311\}$ and BCC- $\{211\}$ and $\{321\}$ crystallographic plane families along the transverse direction.....	16
4. Dislocation density calculation.....	17
Supplementary Fig. 12 Modified Williamson-Hall plots obtained from the BCC and FCC peaks in the <i>in situ</i> neutron diffraction patterns of the as-printed AlCoCrFeNi _{2.1} EHEA.....	17
Supplementary Table 3 Evolution of dislocation densities in the BCC and FCC phases of as-printed AlCoCrFeNi _{2.1} EHEA upon tensile deformation.	18
5. Lattice mismatch evolution	18
Supplementary Table 4 Evolution of lattice mismatch between BCC and FCC phases in as-printed AlCoCrFeNi _{2.1} EHEA upon tensile deformation.	18

S4 POST HEAT TREATMENT EFFECT	19
Supplementary Fig. 13 Tensile properties of as-printed and annealed AlCoCrFeNi _{2.1} EHEAs.	19
Supplementary Fig. 14 Phase equilibria and composition predicted by CALPHAD.	20
Supplementary Fig. 15 Phase evolutions of AM AlCoCrFeNi _{2.1} EHEA after different heat treatments measured by high-energy X-ray diffraction.	21
Supplementary Fig. 16 Microstructure evolutions of AM AlCoCrFeNi _{2.1} EHEAs after different heat treatments.	23
S5 SUMMARY OF TENSILE PROPERTIES OF ADDITIVELY MANUFACTURED HIGH-PERFORMANCE METAL ALLOYS.....	24
Supplementary Table 5 A summary of tensile properties, including yield strength ($\sigma_{0.2}$), ultimate tensile strength (σ_u), uniform elongation (ε_u), and fracture elongation (ε_f) of the high-strength metal alloys processed by additive manufacturing	24
S6 DP-FE MODELLING AND SIMULATIONS	26
1. Micromechanics model	26
Supplementary Table 6 Elastic constants of FCC and BCC phases in the AlCoCrFeNi _{2.1} EHEA determined from the micromechanics model.	28
2. DP-CPFE model	28
Supplementary Table 7 Parameters used in the DP-CPFE model	29
Supplementary Fig. 17 DP-CPFE comparison between the average stress-strain response of a colony/grain containing an FCC/BCC lamellar structure and the average stress-strain response of an FCC and a BCC colony/grain, as embedded in the same homogenized poly-colony/grain matrix under an applied tensile stress σ_x^a	31
S7 DESCRIPTION OF <i>IN SITU</i> TEM VIDEO	32
Supplementary Video 1: <i>In situ</i> TEM showing dislocation movements in BCC and FCC nanolamellae under tensile loading	32
Supplementary Fig. 18 Deforming nanolamellar structure and schematic illustration in an <i>in situ</i> TEM straining experiment (Supplementary Video 1).	32
S8 <i>IN SITU</i> THERMAL CYCLING EFFECT	33
Supplementary Fig. 19 Secondary electron images showing the microstructure in an as-printed single-track bead and a multi-layer bulk sample.	34
Supplementary Fig. 20 Characterization of microstructures and mechanical properties along the build direction of as-printed AlCoCrFeNi _{2.1} thin-wall structures.	35
Supplementary Fig. 21 Microstructure of as-printed AlCoCrFeNi _{2.1} octet lattices..	36
Supplementary References.....	37

S1 OPTIMIZATION OF PROCESSING PARAMETERS

To enable the optimization of laser processing parameters, a normalized processing diagram for AlCoCrFeNi_{2.1} EHEA is constructed to help understand the relationship between 3D printing processing parameters, microstructures, and mechanical properties. In this diagram, two dimensionless processing variables $E_{\min}^*$ and l/h^* are used as the x and y axis. Here, we define¹

$$E_{\min}^* = \frac{p^*}{v^* l^*} = \frac{Ap}{2vlr_b \rho C_p (T_m - T_0)} \text{ and } h^* = \frac{h}{r_b}$$

where $E_{\min}^*$ is the normalized minimum volumetric energy density required to melt a material; p^* , v^* , l^* and h^* are the dimensionless laser power, scan speed, layer thickness and hatch spacing, respectively; A is the surface absorptivity (~ 0.35), p the laser power, v the scan speed, l the layer thickness, h the hatch spacing, r_b the laser beam radius (5×10^{-5} m), ρ the density of AlCoCrFeNi_{2.1} (7420 kg/m^3), C_p the specific heat capacity ($\sim 507 \text{ J/kg/K}$, estimated by using the weight averaging method), T_m the melting temperature of AlCoCrFeNi_{2.1} (1616 K , measured by differential thermal analysis), and T_0 the powder bed temperature (353 K). The dimensionless laser power p^* is calculated as

$$p^* = \frac{Ap}{r_b k (T_m - T_0)}$$

where k is the thermal conductivity. The dimensionless scan speed is given by

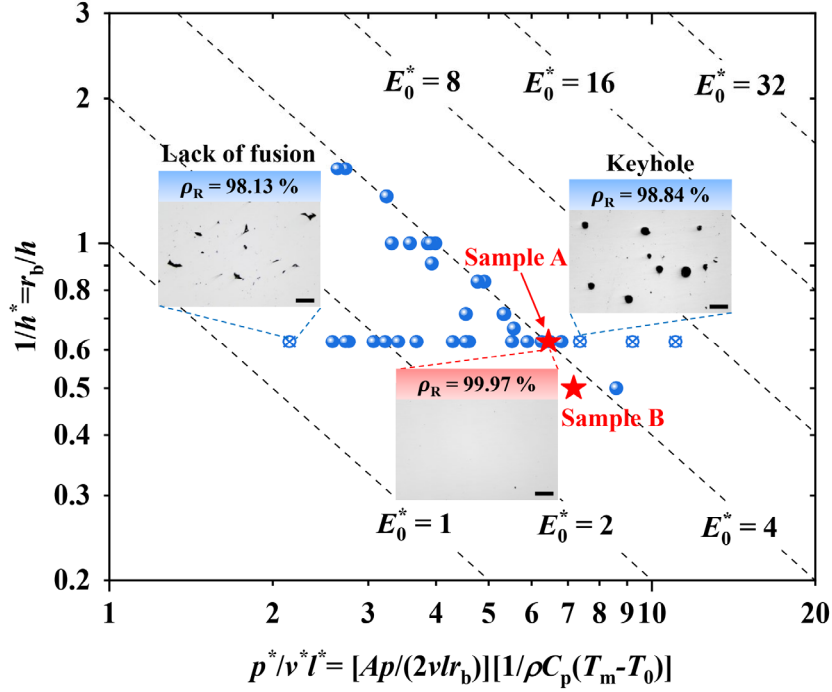
$$v^* = \frac{vr_b}{D}$$

where D is the diffusion coefficient, which can be related to the thermal conductivity and specific heat capacity by the following relation

$$D = \frac{k}{\rho C_p}$$

The dimensionless layer thickness is defined as

$$l^* = \frac{2l}{r_b}$$



Supplementary Fig. 1 | Normalized processing diagram for L-PBF of AlCoCrFeNi_{2.1}. Isopleths of normalized equivalent energy density (E_0^*) that suggest an equally energy efficient process are presented in dashed lines. Optimized processing conditions used in this work are highlighted by red stars. The processing conditions that yield samples with the relative density (ρ_R) above or below 99.5% are depleted as solid and open circles, respectively. The insets show the representative optical microscopy images of the cross-section of the samples falling into the lack-of-fusion, fully-dense, and keyhole regions, respectively. Scale bar, 100 μm .

Using the normalized equivalent energy density method¹, we identified an effective laser processing window to print fully-dense EHEA samples of AlCoCrFeNi_{2.1} (Supplementary Fig. 1). The E_0^* diagram provides a ready means in delineating the convoluted processing parameters. We observed that near fully dense ($> 99.5\%$) samples fall in the range of $1.5 \leq E_0^* \leq 4$, while $E_0^* < 1.5$ leads to lack-of-fusion and $E_0^* > 4$ triggers keyholing. At a specific optimal E_0^* , increasing the cooling rate yields a smaller interlamellar spacing for the AM EHEA, as shown in the two representative cases denoted as sample A and sample B (Supplementary Table 1 and Supplementary Fig. 2). Despite their microstructural similarity, we focused on characterizing sample A with finer interlamellar spacings and used sample B to demonstrate the tunability of nanolamellar structures and associated properties.

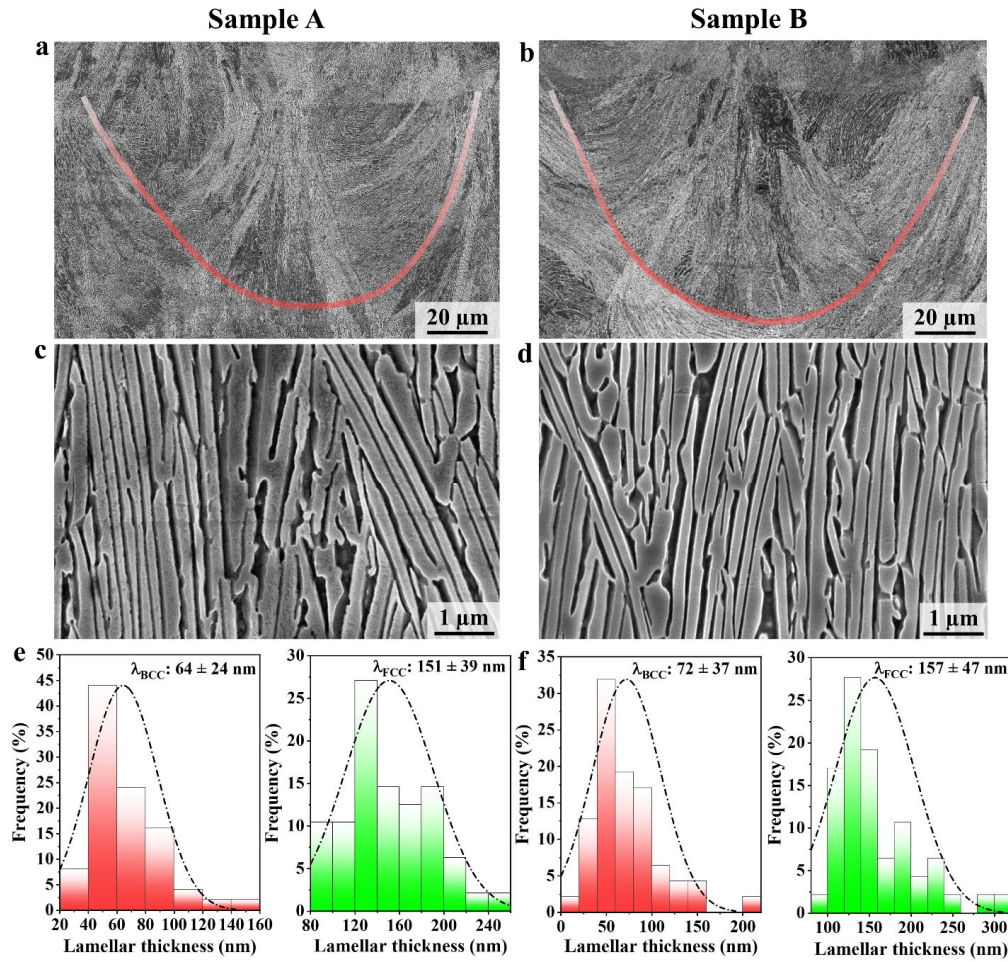
Considering the heat balance during solidification and the Jackson-Hunt model² that correlates the interlamellar spacing, λ , to the solidification rate, u , by a constant as $K = \lambda^2 u$, the cooling rate dependent interlamellar spacing is expressed as $\lambda^2 = \frac{\Delta H_f}{V_m C_p} \frac{3K}{R(dT/dt)}$, where C_p is the heat capacity,

dT/dt the cooling rate, ΔH_f the enthalpy of fusion, V_m the molar volume and R the radius of the hemispherical melt pool³. The average melt pool dimension in sample A (width: $\sim 131 \mu\text{m}$, depth: $\sim 69 \mu\text{m}$) is similar to that in sample B (width: $\sim 140 \mu\text{m}$, depth: $\sim 78 \mu\text{m}$). However, subject to a higher laser power and a lower laser scan speed, sample B underwent a lower cooling rate (dT/dt

$= G \times u$), which is originated from a smaller thermal gradient (G) and a slower solidification rate (u). Consequently, sample A with the higher cooling rate is characteristic of a smaller interlamellar spacing ($\lambda \approx 215$ nm) as compared with sample B ($\lambda \approx 229$ nm).

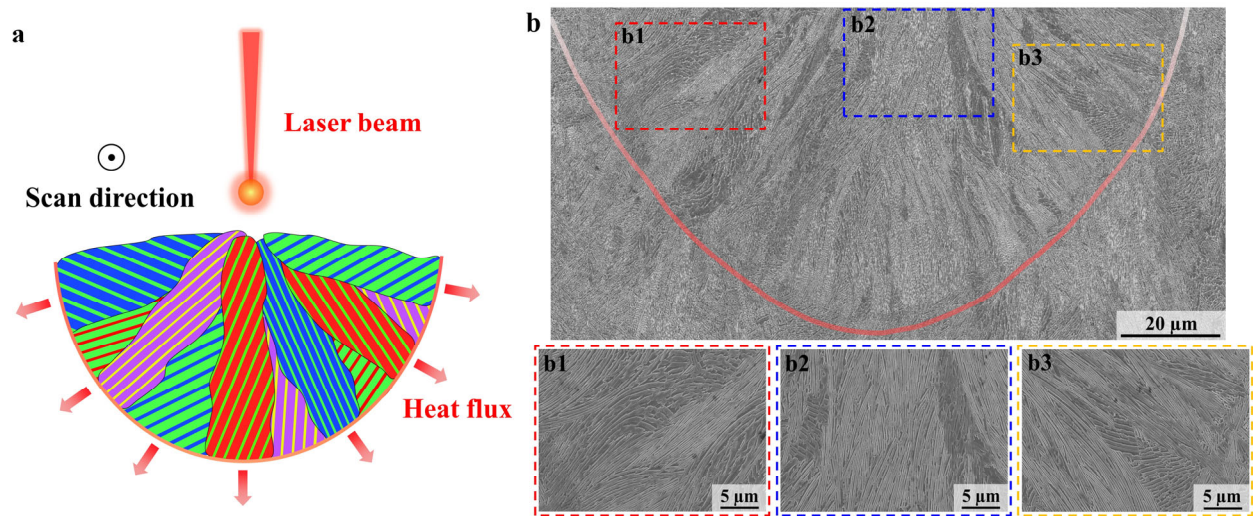
Supplementary Table 1 | L-PBF processing parameters for optimized EHEA samples.

Processing parameters	Sample A	Sample B
Laser power (W)	350	370
Scan speed (mm/s)	1000	950
Layer thickness (μm)	40	40
Hatch spacing (μm)	80	100
Scan pattern	bi-directional	bi-directional
Rotation angle ($^\circ$)	90	90

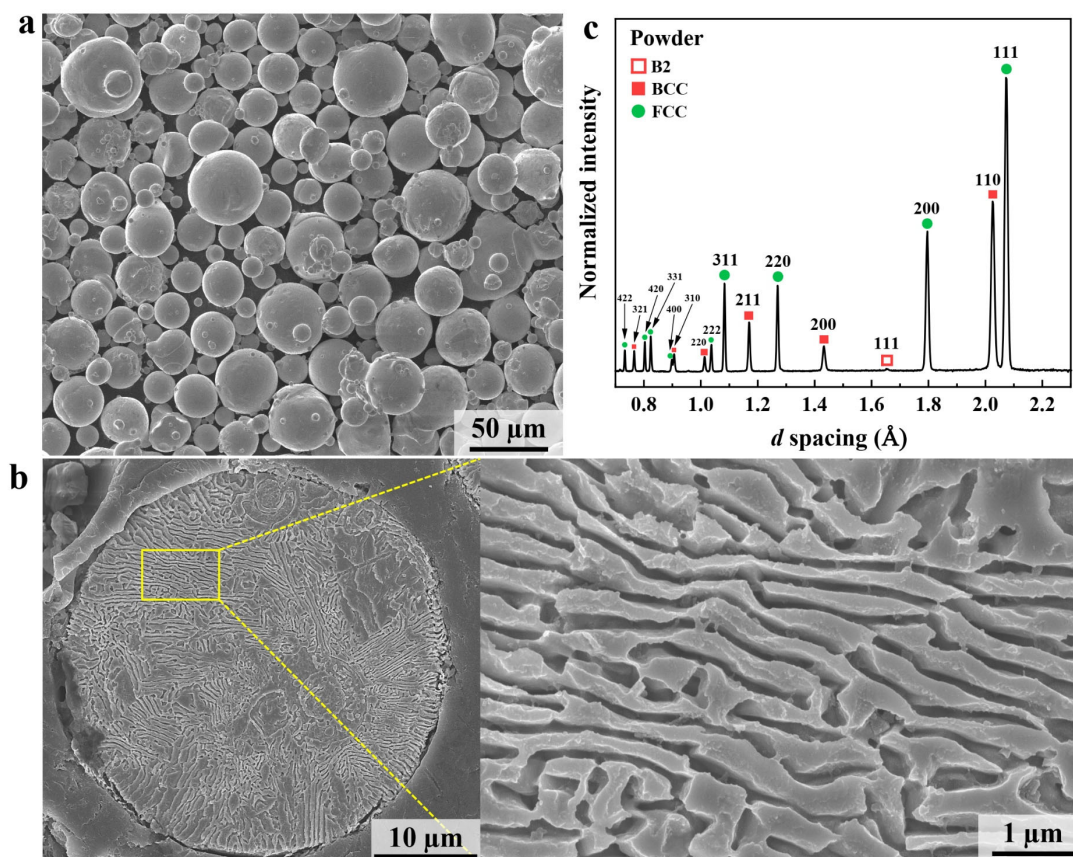


Supplementary Fig. 2 | Secondary electron images of the cross-section views of the optimized samples A and B. a and b, The red lines highlight the near-hemispherical melt pool boundaries in the optimized samples A and B. **c and d,** Representative nanolamellar structures of the optimized samples A and B, taken from the centre of the melt pools. **e and f,** Histograms showing the BCC and FCC lamellar thickness distributions in samples A and B based on statistical analysis of multiple secondary electron images.

S2 MICROSTRUCTURAL CHARACTERIZATION



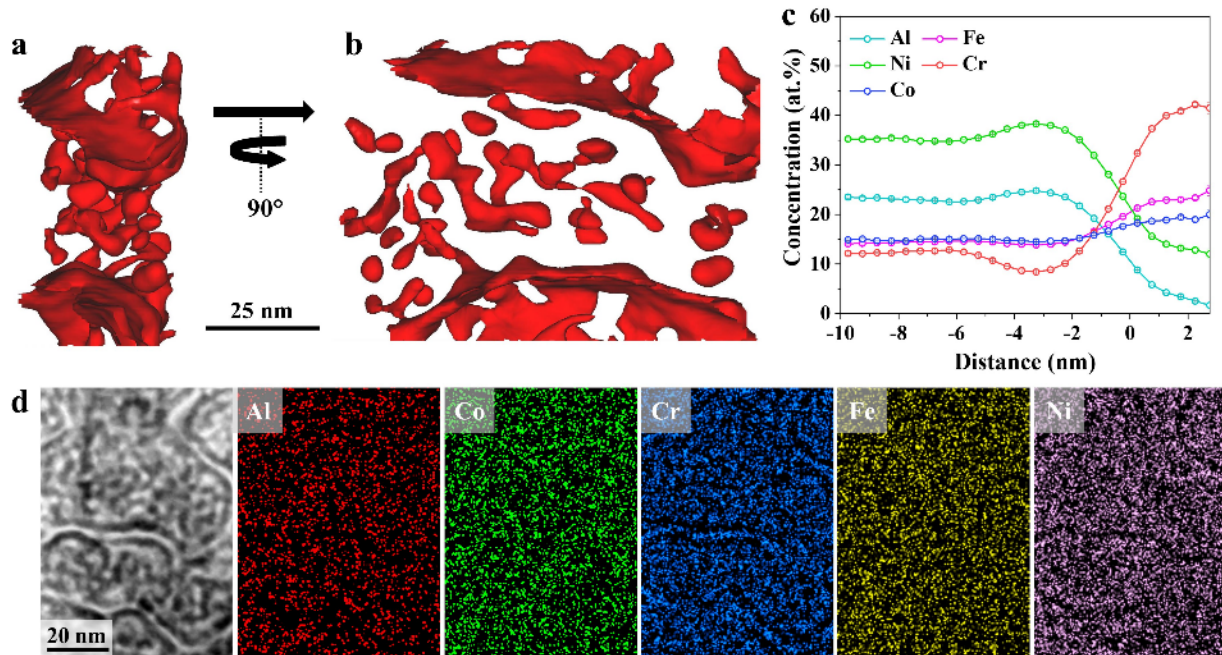
Supplementary Fig. 3 | Nearly random orientations in shape and crystallography of nanolamellar eutectic colonies in AM AlCoCrFeNi_{2.1} EHEA. **a**, Schematic illustration of the nearly random orientations in shape and crystallography of eutectic colonies inside a melt pool. The thick and thin lamellae in each eutectic colony represent the FCC and BCC phases, respectively. The heat flux is perpendicular to the local melt pool boundary, and the elongated eutectic colonies are oriented toward the centre of the melt pool. Small track overlap is not taken into account in the schematic. **b**, Secondary electron image of a cross-section view of the as-printed EHEA, showing the overall random shape orientations of eutectic colonies inside a melt pool. The melt pool boundary is outlined by the red line. **b1-b3**, Magnified images showing regions with different shape orientations of nanolamellar eutectic colonies inside the melt pool.



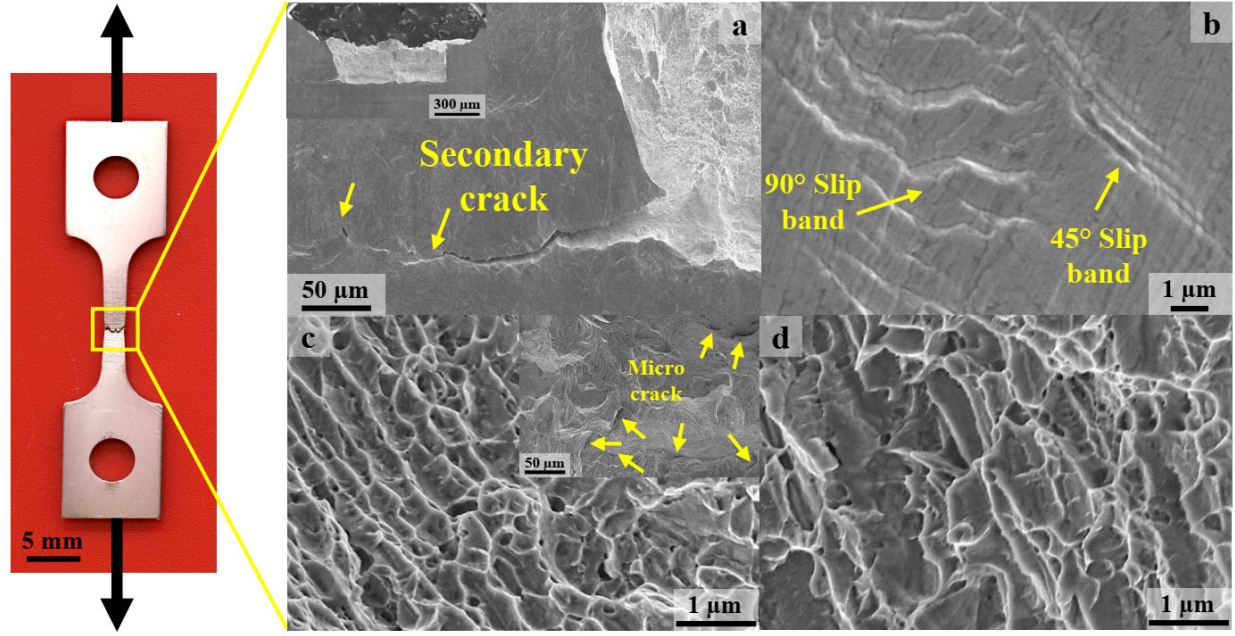
Supplementary Fig. 4 | Microstructural characterization of the AlCoCrFeNi_{2.1} powder feedstock. a, SEM micrograph of the gas-atomized AlCoCrFeNi_{2.1} powders with the particle size of 15-53 μm. **b,** SEM micrograph of the cross-section view of the AlCoCrFeNi_{2.1} powder showing that the pre-alloyed powders are free of defects and characteristic of a lamellar structure with an interlamellar spacing of $\lambda \approx 355$ nm. **c,** Neutron diffraction pattern of the AlCoCrFeNi_{2.1} powders with FCC+BCC/B2 phases.

Supplementary Table 2 | Chemical compositions of AlCoCrFeNi_{2.1} powders and as-printed sample (wt.%) measured by ICP-MS and IGA.

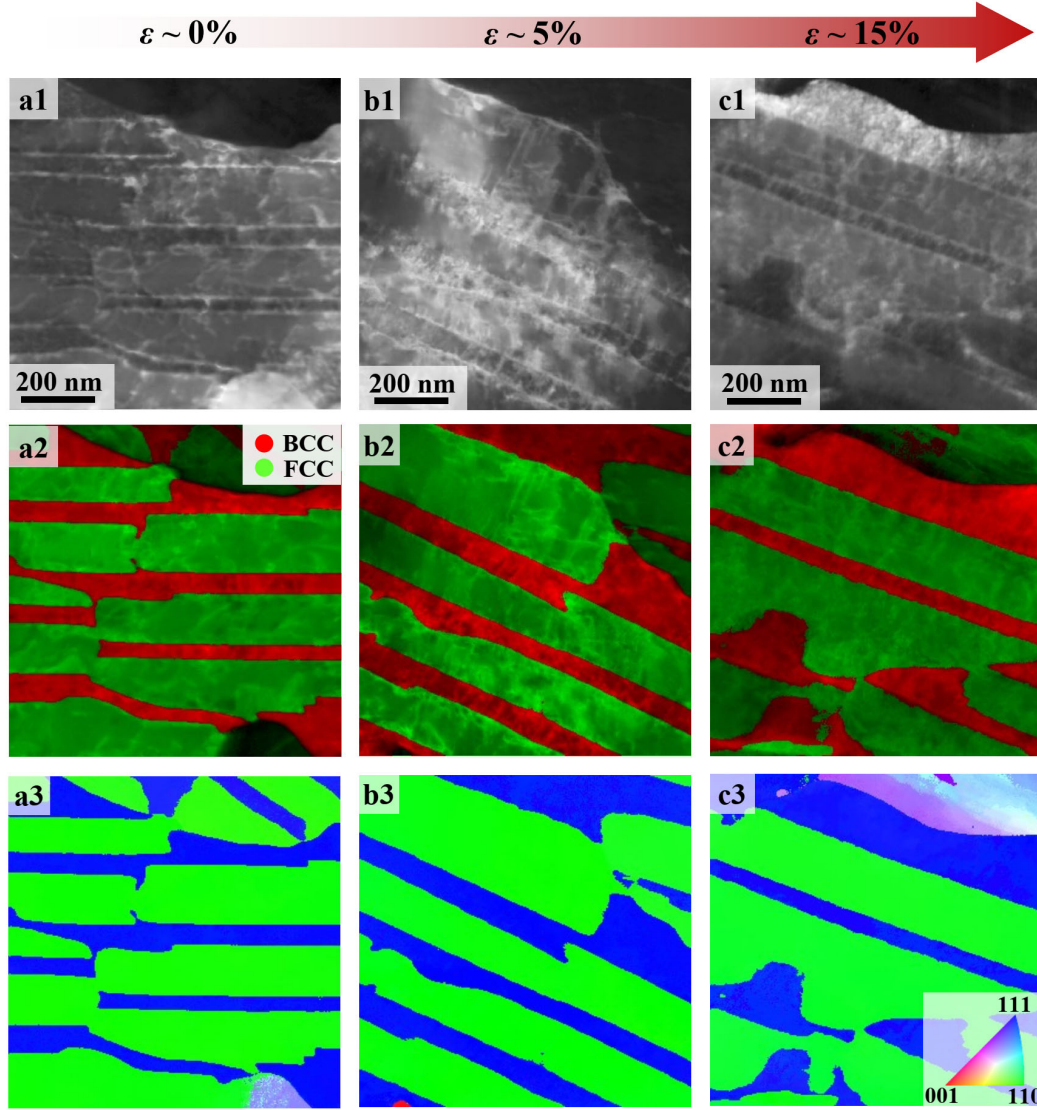
	Al	Co	Cr	Fe	Ni	O	N
Powder feedstock	8.42 ± 0.06	18.0 ± 0.0	16.0 ± 0.1	18.2 ± 0.1	39.4 ± 0.1	-	-
As-printed sample	8.62 ± 0.08	17.9 ± 0.0	15.8 ± 0.0	18.3 ± 0.1	39.4 ± 0.0	0.0031 ± 0.0002	0.012 ± 0.002



Supplementary Fig. 5 | Microstructural characterization of the chemically-modulated nanostructures within BCC lamellae of the as-printed AlCoCrFeNi_{2.1} EHEA. **a, The bi-continuous chemically-modulated nanostructures in BCC lamellae are highlighted by the isoconcentration surfaces encompassing regions containing more than 27 at. % of Cr. **b**, The isoconcentration surfaces displayed in **a** rotated by 90° along the long axis. **c**, APT proximity histogram of the 27 at. % Cr-isoconcentration surfaces shown in **a**. **d**, HAADF-STEM image and the corresponding STEM-EDX mapping of all constituent elements within the chemically-modulated BCC lamellae, manifested by the nanoscale Ni-Al-rich and Cr-rich regions.**



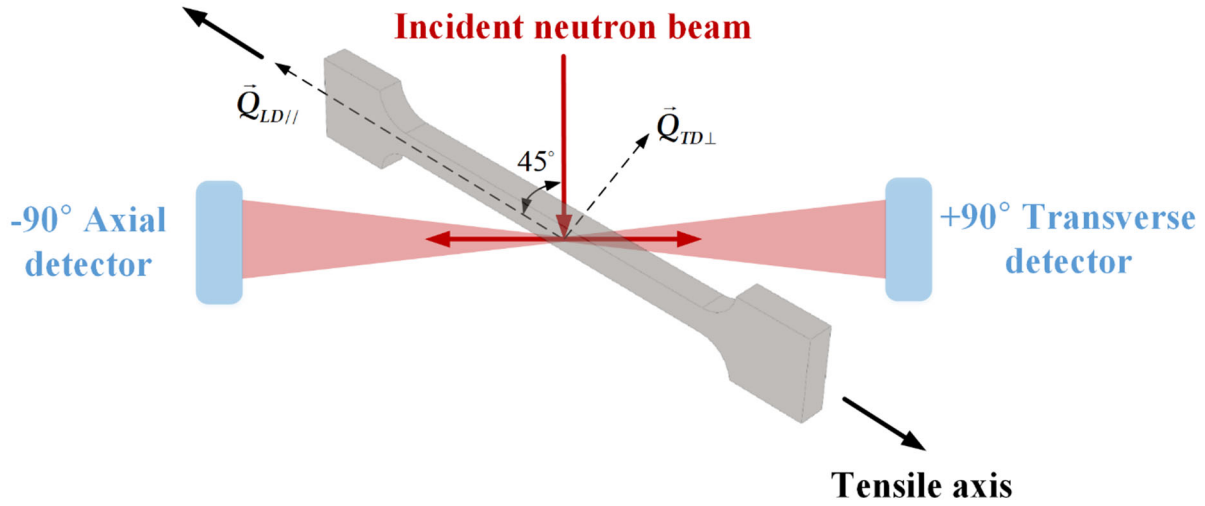
Supplementary Fig. 6 | Fracture morphologies of the as-printed AlCoCrFeNi_{2.1} EHEA. **a** and **b** display the continuous wavy slip traces on the side surface near the fracture end, indicating the co-deformation of the FCC and BCC lamellae. Secondary cracks were observed to be aligned with the laser scan direction. The upper-left inset of **a** shows the “cup-and-cone” shaped fracture surface, which is typical for ductile fracture. **c** and **d** show the fracture surface morphologies of the as-printed sample decorated with microcracks and dimples. No trench-like features as reported in the as-cast sample⁴ were observed, suggesting the compatible co-deformation of the BCC and FCC phases in the AM AlCoCrFeNi_{2.1}.



Supplementary Fig. 7 | Meso-scale deformation structures of the as-printed AlCoCrFeNi_{2.1} EHEA imaged by PED. a1-c1, Virtual dark-field PED micrographs revealing the evolutions of dislocation substructures in BCC and FCC nanolamellae at different tensile strains. **a2-c2**, Corresponding phase maps of **a1-c1**. **a3-c3**, Corresponding inverse-pole figure (IPF) maps of **a1-c1** acquired by PED-based automated crystal orientation mapping (ACOM) in TEM, indicating the maintenance of K-S orientation relationship between the BCC and FCC phases over the entire deformation.

S3 *IN SITU* NEUTRON DIFFRACTION MEASUREMENTS UNDER TENSION

1. Experimental set-up



Supplementary Fig. 8 | Schematic diagram of *in situ* neutron diffraction tensile testing set-up.

2. Determination of reference $d_{0,hkl}$ and residual stress

The extreme processing conditions (e.g., large thermal gradients and rapid cooling rate) of L-PBF tend to induce residual stresses in AM alloys. In general, the residual stresses in polycrystalline materials can be categorized into two major classes: macroscale (type I) and microscale (type II) residual stresses^{5,6}. The macroscale (type I) residual stresses are distributed across the overall dimension of a printed part. Several review articles in literature^{7,8} have discussed the origin and characteristics of macroscale residual stresses in AM alloys. Measurement of macroscale residual stresses is often conducted for a printed part on the substrate. In our work, the macroscale residual stresses are likely to be largely released after we cut the samples from the substrate for microstructure characterization and mechanical testing. After sample cutting, we did not observe obvious distortion, which indicates that the macroscale residual stresses in our AM EHEA are not significant.

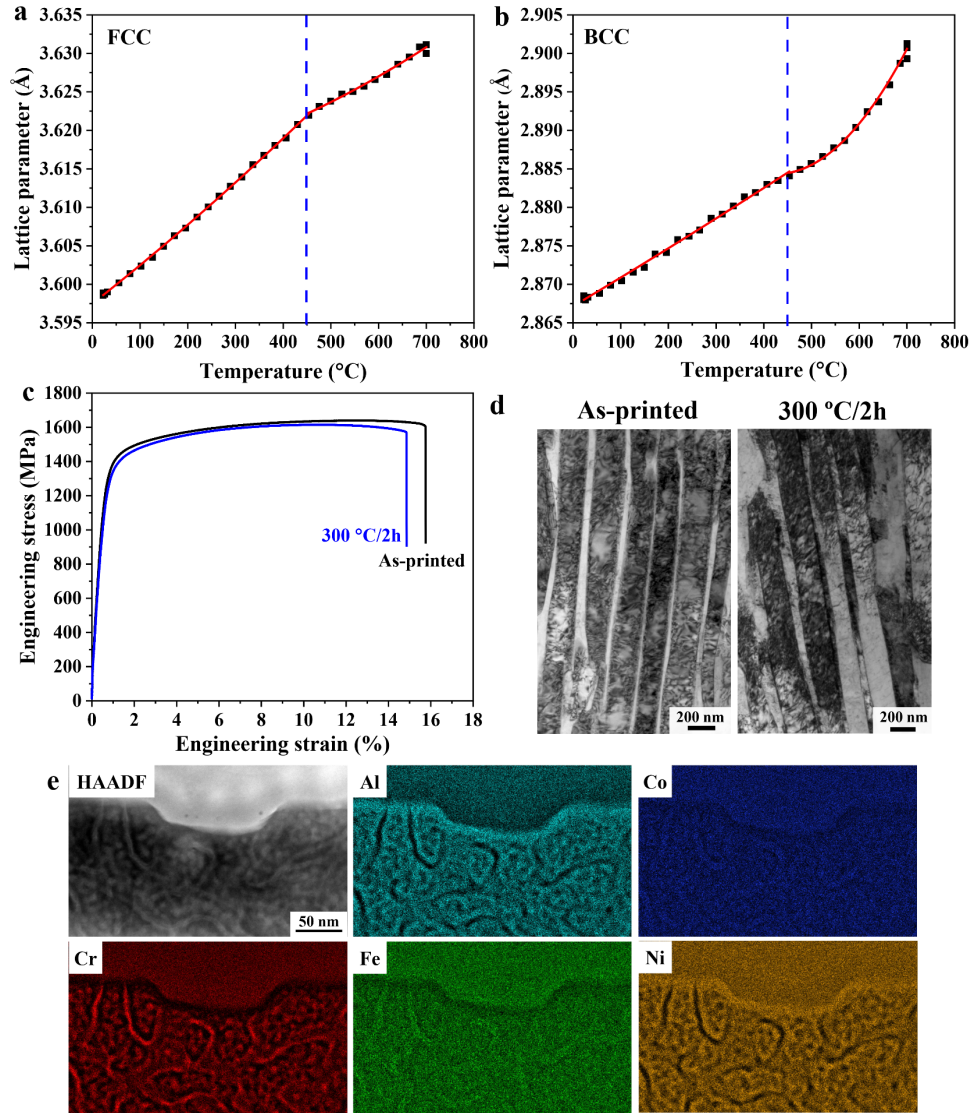
While most efforts to evaluate residual stresses in AM materials were devoted to the macroscale (type I) residual stresses⁷, few studies have been performed to investigate the microscale (type II) residual stresses at the microstructure level⁹. The type II residual stresses are internal stresses that are self-equilibrated over the length scale of grains and usually result from strain incompatibility between grains in a printed sample after cooling. Next, we report our studies of type II intergranular/interphase residual stresses in the AM EHEA.

Specifically, we performed additional neutron diffraction experiments to evaluate the type II intergranular/interphase residual stresses in the AM EHEA. Note that a “stress-free” state needs to be determined for evaluation of the lattice strains and thus microscale residual stresses in different $\{hkl\}$ reflections. To this end, we performed *in situ* heating (at a rate of ~20 °C/min) neutron diffraction experiment. As shown in Supplementary Fig. 9a and b, the lattice parameters of both FCC and BCC phases increased linearly with temperature below 450 °C. After heating above 450 °C, the increase of lattice parameter with temperature started to deviate from linearity, indicating

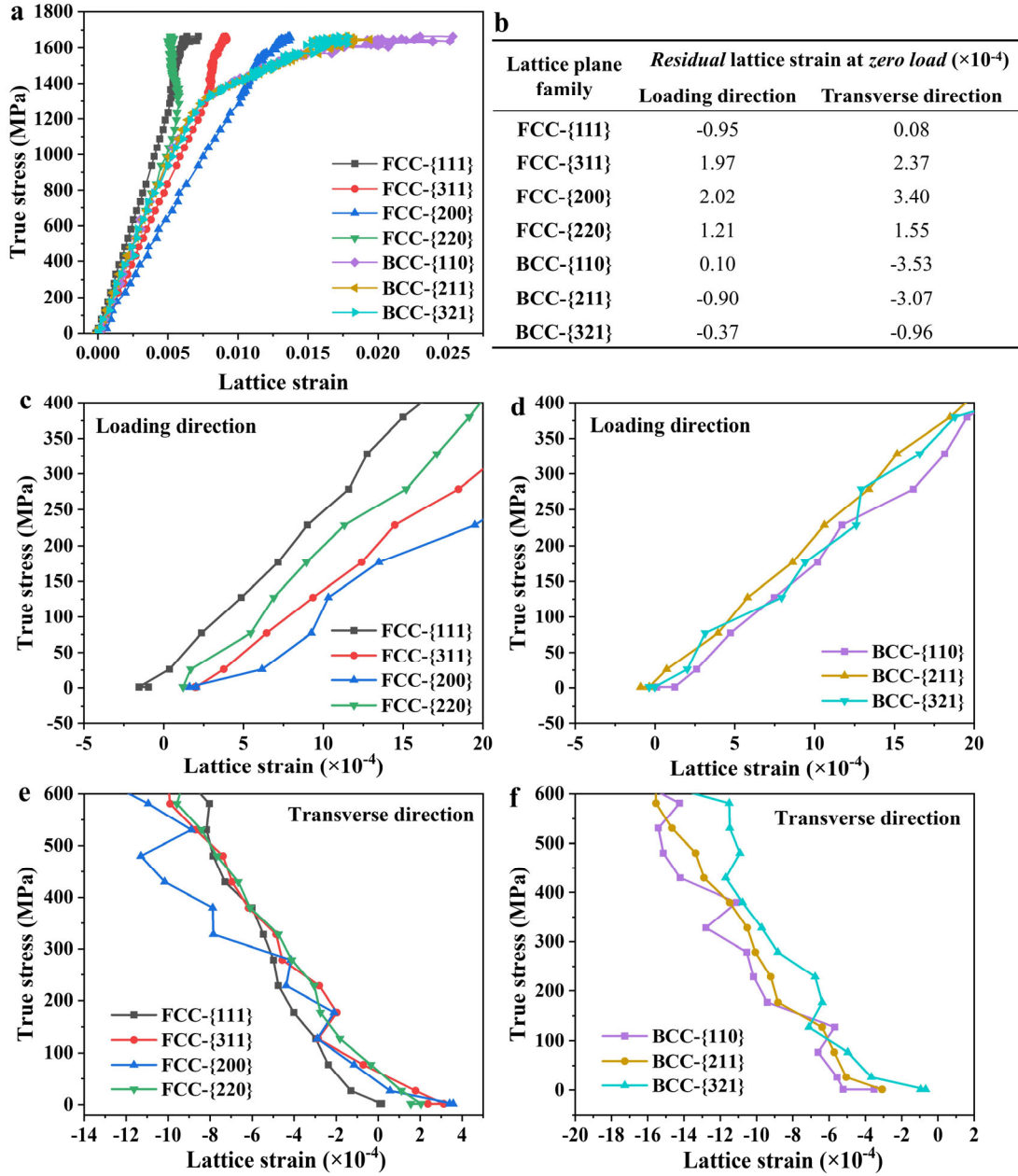
the possible compositional changes in one or both phases. Based on these results, we performed low-temperature annealing to relieve the type II intergranular/interphase residual stresses, while hoping to retain the main microstructure characteristics of the AM EHEA. We further found that the sample annealed at 300 °C/2h could provide a desired “stress-free” reference. This is evidenced by the tensile response of the 300 °C/2h-annealed sample that is close to that of the as-printed sample (Supplementary Fig. 9c). In addition, the TEM micrograph in Supplementary Fig. 9d shows a similar nanolamellar structure of the 300 °C/2h-annealed sample as the as-printed sample. The characteristic nanoscale chemical modulation, as manifested by the intertwined Ni-Al rich and Co-Cr-Fe rich regions within the BCC nanolamellae, was also retained after annealing (Supplementary Fig. 9e). From the measured d spacings of different $\{hkl\}$ reflections of the 300 °C/2h-annealed sample, we obtained the “stress-free” reference values of $d_{0,hkl}$.

Next, we used *in situ* neutron diffraction tensile testing results of the as-printed sample to calculate the lattice strains of different $\{hkl\}$ reflections by $\varepsilon_{hkl} = (d_{hkl} - d_{0,hkl})/d_{0,hkl}$ (where d_{hkl} denotes the measured d spacings) as shown in Supplementary Fig. 10a. From these calculations, we extracted the *residual* lattice strains of different $\{hkl\}$ reflections in each phase of the as-printed sample, as listed in Supplementary Fig. 10b and also plotted in Supplementary Fig. 10c-f. It is seen that in our as-printed EHEA, the BCC lamellae exhibit compressive residual strains while the FCC lamellae exhibit tensile residual strains, both of which are on the order of $10^{-4} \sim 10^{-5}$. The corresponding type II intergranular/interphase residual stresses (as evaluated from generalized Hooke’s law) are negligibly small compared to the typical yield and flow strengths achieved by the AM EHEA. These residual stresses are mainly of the hydrostatic type, which likely stems from a small mismatch of the coefficient of thermal expansion (CTE) between the dual phases.

Finally, we note that the microscale residual stresses are inevitably present in the dual-phase materials even after annealing. This is due to the intrinsic CTE mismatch between the dual phases. Nonetheless, annealing can partially remove the microscale residual stresses developed during L-PBF processing. Hence, the low-temperature annealed condition we adopted facilitates the attainment of a relatively “stress-free” state.

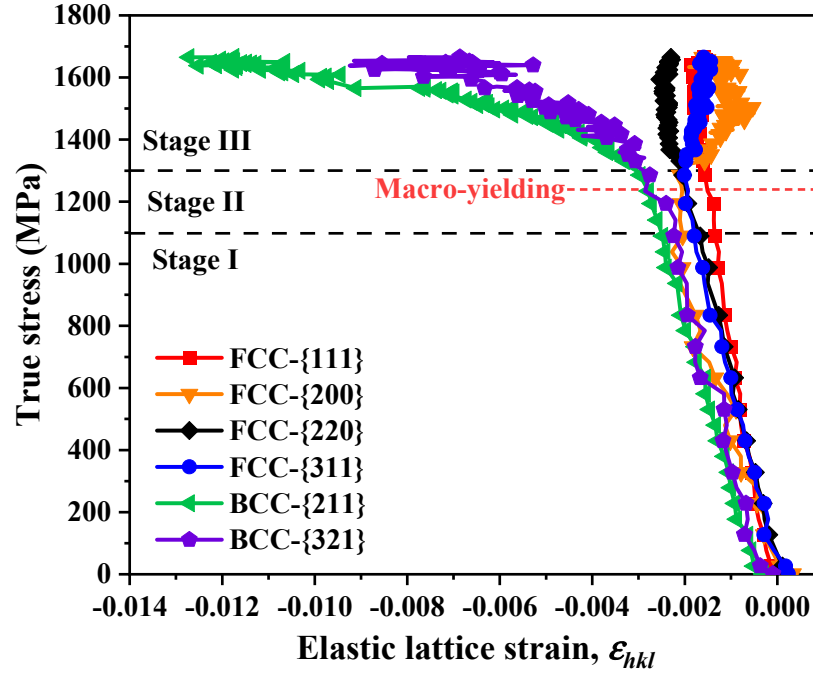


Supplementary Fig. 9 | Determination of a “stress-free” AM AlCoCrFeNi_{2.1} EHEA sample through thermal annealing. **a**, The lattice parameter of FCC phase varies with temperature, from *in situ* neutron diffraction measurement with continuous heating. **b**, Same as **a**, except for BCC phase. **c**, Tensile stress-strain curves of the as-printed and 300 °C/2h-annealed AlCoCrFeNi_{2.1} EHEA samples. **d**, Bright-field TEM images of the as-printed and 300 °C/2h-annealed AlCoCrFeNi_{2.1} EHEA samples. **e**, HAADF STEM-EDX maps showing chemical fluctuations manifested by the nanoscale Ni-Al-rich and Co-Cr-Fe-rich regions inside BCC lamellae of the 300 °C/2h-annealed sample.



Supplementary Fig. 10 | Characterization of microscale residual stresses through *in situ* neutron diffraction measurement of lattice strains in AM AlCoCrFeNi_{2.1} EHEA. a, Evolution of lattice strain against macroscopic true stress for representative FCC ({111}, {200}, {220}, {311}) and BCC ({110}, {211} and {321}) lattice plane families along the loading direction. The stress-free lattice spacing $d_{0,hkl}$ was determined by annealing an as-printed sample at 300 °C for 2 h. **b**, Tabulated *residual* lattice strains (without applied loading) for different lattice plane families along loading and transverse directions. **c**, Enlarged curves in **a** under small applied tensile strains, showing the true stress versus lattice strain responses along the loading direction for FCC-{111}, {200}, {220}, {311} lattice plane families. **d**, Enlarged curves in **a** under small applied tensile strains, showing the true stress versus lattice strain responses along the loading direction for BCC-{110}, {211}, {321} lattice plane families. **e**, Same as **c**, except for lattice strain responses along the transverse direction. **f**, Same as **d**, except for lattice strain responses along the transverse direction.

3. Lattice strain evolution along transverse direction



Supplementary Fig. 11 | Evolution of the orientation-specific elastic lattice strain over macroscopic true stress for FCC-{111}, {200}, {220}, {311} and BCC-{211} and {321} crystallographic plane families along the transverse direction. The 0.2% offset yield strength is marked by the red dashed line.

4. Dislocation density calculation

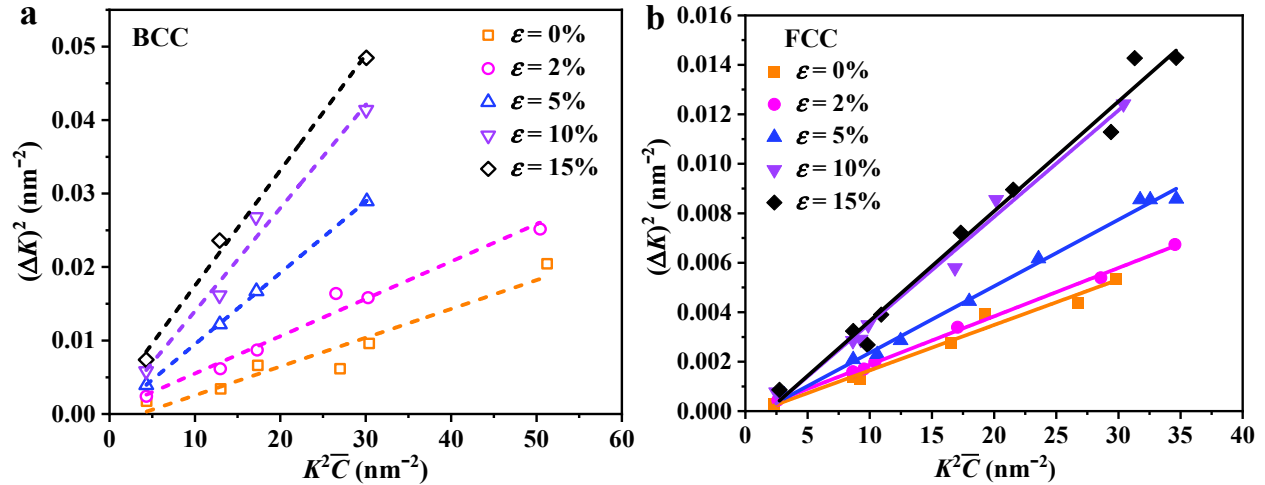
The dislocation density was calculated by the modified Williamson-Hall (MWH) method¹⁰ which quantifies the effects of the average sub-grain (crystallite) size D and the micro-strain fields on line broadening as follows:

$$(\Delta K)^2 = (0.9/D)^2 + (\pi A^2 b^2 \rho / 2) (K^2 \bar{C}) + O(K^2 \bar{C})^2$$

Here $K = 1/d$, $\Delta K = -K(\Delta d/d)$, d is the interplanar spacing and Δd is the full width at half maximum (FWHM) of the diffraction peaks, b the Burgers vector length of dislocation, ρ the dislocation density, A a constant depending on the effective outer cutoff radius of dislocations, and O representing the non-interpreted high order terms of $K^2 \bar{C}$. Strain anisotropy among each hkl reflection is rationalized by the dislocation contrast factor C_{hkl} . Under the assumption that all possible slip systems are equally populated, the average dislocation contrast factor, $\bar{C}$, can be determined by:

$$\bar{C} = \bar{C}_{h00} (1 - qH^2)$$

where $H^2 = (h^2 k^2 + k^2 l^2 + h^2 l^2) / (h^2 + k^2 + l^2)^2$, and $\bar{C}_{h00}$ is a constant depending on the dislocation type and anisotropic elastic constants C_{11} , C_{12} , and C_{44} . q is a variable related to the fractions of screw and edge dislocations, which can be experimentally determined by the linear regression of $[(\Delta K)^2 - (0.9/D)^2]/K^2$ versus H^2 . Accordingly, the slope (m) of the MWH plot, $(\Delta K)^2$ versus $K^2 \bar{C}$ (Supplementary Fig. 12), gives the product $\pi A^2 b^2 \rho / 2$. The error of the dislocation densities was quantified based on the uncertainties of the slope m from linear regression. To eliminate the effect of stress fields which vary with grains on line broadening, the FWHM extracted from the diffraction patterns when the sample was unloaded to near-zero load was used to calculate the dislocation density. The evolution of dislocation densities in BCC and FCC phases at different tensile strains are listed in Supplementary Table 3.



Supplementary Fig. 12 | Modified Williamson-Hall plots obtained from the BCC (a) and FCC (b) peaks in the *in situ* neutron diffraction patterns of the as-printed AlCoCrFeNi_{2.1} EHEA.

Supplementary Table 3 | Evolution of dislocation densities in the BCC and FCC phases of as-printed AlCoCrFeNi_{2.1} EHEA upon tensile deformation.

Strain	Dislocation density (/m ²)	
	FCC (ρ_{FCC})	BCC (ρ_{BCC})
0	$5.4 \times 10^{14} \pm 0.3 \times 10^{14}$	$7.4 \times 10^{14} \pm 1.1 \times 10^{14}$
2%	$5.7 \times 10^{14} \pm 0.1 \times 10^{14}$	$9.5 \times 10^{14} \pm 0.7 \times 10^{14}$
5%	$7.9 \times 10^{14} \pm 0.2 \times 10^{14}$	$1.81 \times 10^{15} \pm 0.02 \times 10^{15}$
10%	$1.27 \times 10^{15} \pm 0.06 \times 10^{15}$	$2.61 \times 10^{15} \pm 0.24 \times 10^{15}$
15%	$1.31 \times 10^{15} \pm 0.07 \times 10^{15}$	$2.93 \times 10^{15} \pm 0.20 \times 10^{15}$

5. Lattice mismatch evolution

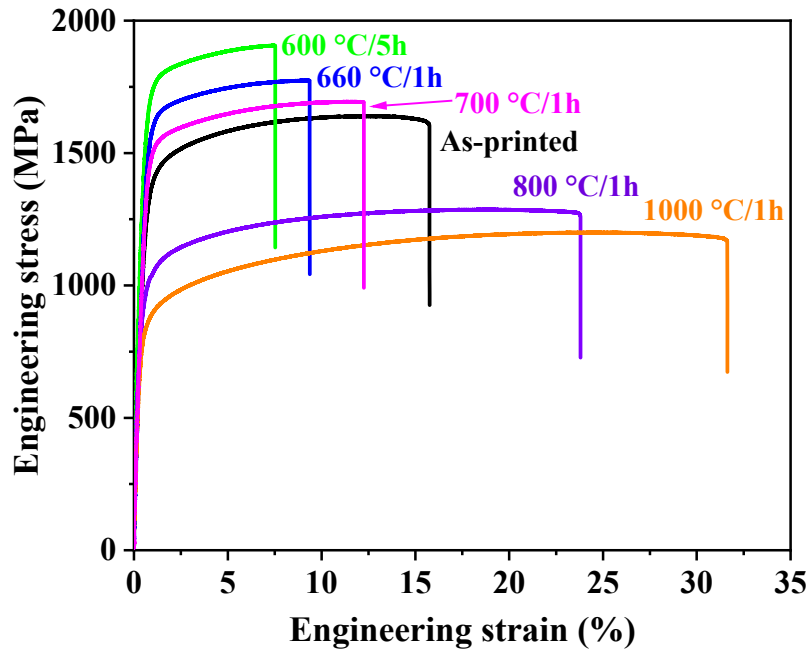
The lattice mismatch between the BCC and the FCC phases is estimated by the relation $\delta = 2(a_{\text{FCC}} - a_{\text{BCC}}/\cos\theta) / (a_{\text{FCC}} + a_{\text{BCC}}/\cos\theta)$, where a is the lattice parameter of each phase and θ is the angle between $\{111\}$ and $\{110\}$ planes. The calculated lattice mismatch at different tensile strains is listed in Supplementary Table 4. The monotonically decreasing lattice mismatch and persistent K-S orientation relationship between the two phases during deformation (Supplementary Fig. 7) are conducive to slip transmission across the semi-coherent phase interfaces.

Supplementary Table 4 | Evolution of lattice mismatch between BCC and FCC phases in as-printed AlCoCrFeNi_{2.1} EHEA upon tensile deformation.

Strain	Lattice parameter (Å)		Lattice mismatch (%)
	FCC	BCC	
0	3.596	2.868	2.29
2%	3.615	2.894	1.95
5%	3.617	2.907	1.52
10%	3.619	2.933	0.68

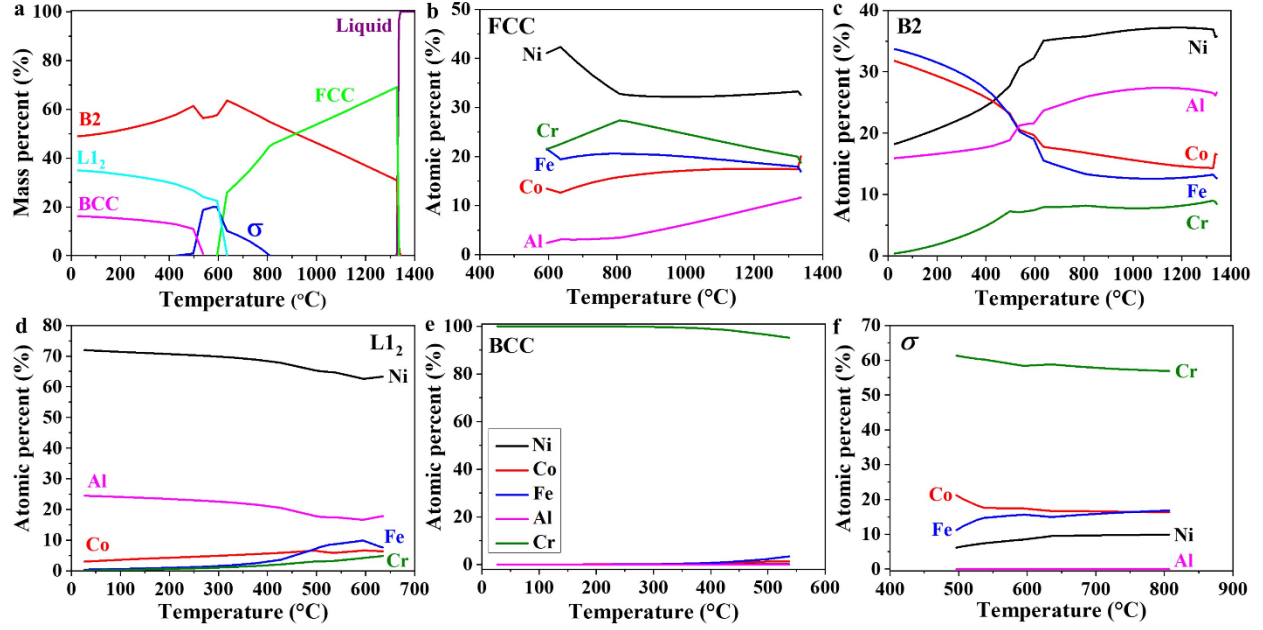
S4 POST HEAT TREATMENT EFFECT

The far-from-equilibrium microstructure of our AM nanolamellae EHEA with dual-phase supersaturated solid solutions not only provides the high strength-ductility combination in the as-printed state, but also offers abundant opportunities to expand the microstructure and material design space by post heat treatment (Fig. 2b and Supplementary Fig. 13). The as-printed dual-phase extended solid solutions are inherently metastable and the supersaturation can decompose by solid-state phase transformation after thermal annealing. To understand the physical origin of the highly tunable mechanical properties, we analysed the phase and microstructure evolution of the AM AlCoCrFeNi_{2.1} EHEAs after thermal annealing at different conditions including 600 °C/5h, 660 °C/1h, 700 °C/1h, 800 °C/1h, and 1000 °C/1h.



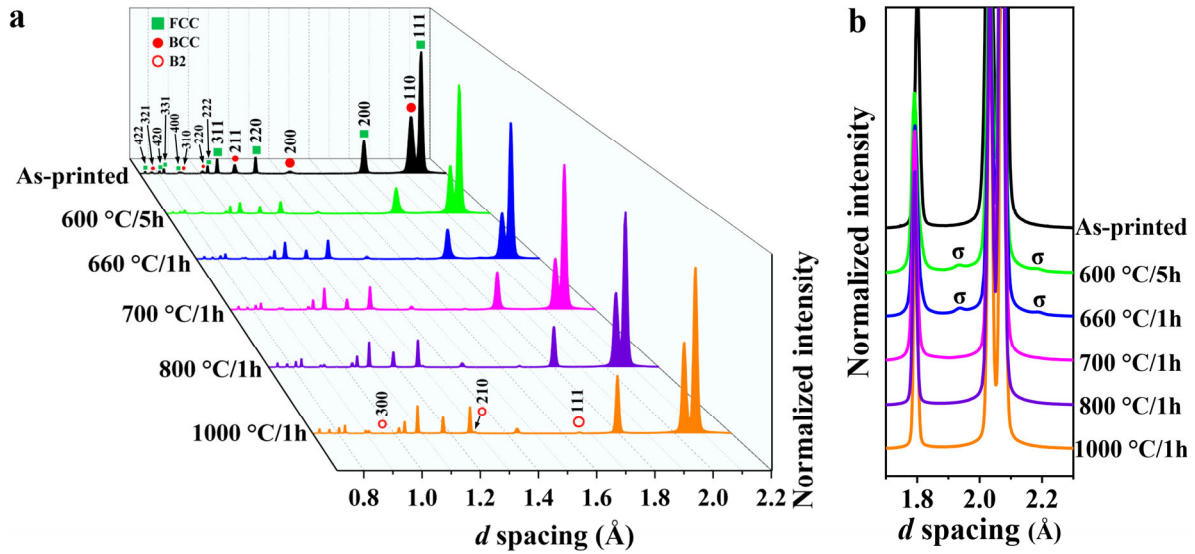
Supplementary Fig. 13 | Tensile properties of as-printed and annealed AlCoCrFeNi_{2.1} EHEAs.

We first consider the thermodynamically equilibrium compositions and phases of the EHEA studied in this work. Although the microstructure after thermal annealing at elevated temperatures may still not be in equilibrium, an equilibrium phase diagram will at least provide insight into possible solid-state transformations at different temperatures. To this end, we performed a thermodynamic calculation of phase diagram (CALPHAD) analysis to predict the phase equilibria and phase compositions across a wide range of temperatures by Thermo-Calc using high-entropy alloy thermodynamic database (v. 2.1.1) (www.thermocalc.com) and the results are shown in Supplementary Fig. 14. Multiple ordered and disordered phases including an intermetallic σ -phase could be present at intermediate temperatures. One salient feature from the equilibrium compositions of different phases is that both the disordered BCC and the ordered σ phases are highly enriched by Cr and severely depleted in Al. By contrast, the ordered B2 phase is Ni-Al-rich, akin to the strong B2-ordering tendency between Ni and Al.



Supplementary Fig. 14 | Phase equilibria and composition predicted by CALPHAD. **a**, Mass percent of equilibrium phases as a function of temperature. **b-f**, Compositions of elements (at. %) in FCC, B2, L1₂, BCC, and σ phases, respectively.

Next, we studied the phase constituents of different annealed samples by high-energy X-ray diffraction (HEXRD) at Beamline 11-ID-C, Advanced Photon Sources of Argonne National Laboratory. The HEXRD patterns of all annealed samples are presented in Supplementary Fig. 15a. Compared to the as-printed sample, several relatively weak superlattice peaks of ordered B2 phase were observed after annealing. As the annealing temperature was increased up to 1000 °C, the normalized intensity of the B2 superlattice peaks became stronger. Of particular interest is that a small amount of σ -phase was also detected in the 600 °C/5h and 660 °C/1h annealed samples, as shown in Supplementary Fig. 15b, which agrees with the corresponding CALPHAD prediction. Nevertheless, the ordered L1₂ phase was not detected at low annealing temperatures, which could be attributed to the insufficient annealing time for attaining equilibrium.



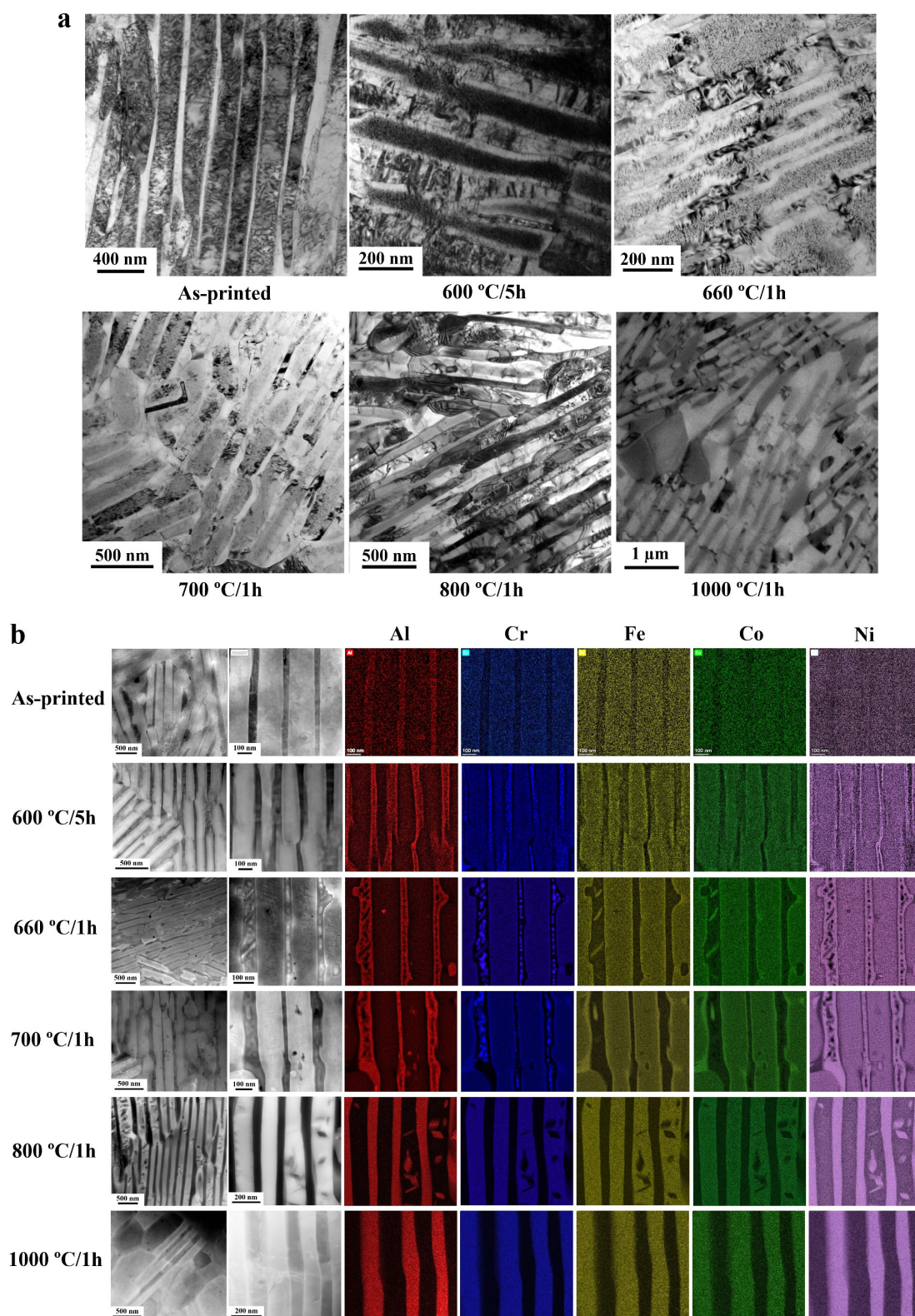
Supplementary Fig. 15 | Phase evolutions of AM AlCoCrFeNi_{2.1} EHEA after different heat treatments measured by high-energy X-ray diffraction (HEXRD). a, HEXRD patterns of as-printed and annealed samples. b, Enlarged diffraction patterns with d spacing ranging from 1.7-2.3 Å, showing the existence of σ phase in the 600 °C/5h and 660 °C/1h annealed samples.

In order to directly observe the microstructural evolution and elemental distribution, we performed additional TEM experiments and elemental mapping by STEM-EDX for all annealed samples. These results are summarized in Supplementary Fig. 16. Overall, an increasingly stronger elemental partitioning between the alternating lamellae was observed as the annealing temperature was increased to promote atomic diffusion (Supplementary Fig. 16b). After annealing at low temperatures (600 °C/5h, 660 °C/1h, 700 °C/1h), Co-Cr-Fe-rich needle-like precipitates and Cr-rich spherical nano-precipitates were found within the BCC lamellae. As a result of such precipitation, the BCC matrix became increasingly rich in Ni-Al, thereby promoting the ordering transformation from BCC to B2 structure. This is evident by the emergence of the relatively weak B2 superlattice peaks in the HEXRD patterns. Moreover, some Ni-Al-rich blocky nano-precipitates were occasionally observed in the FCC lamellae of the 700 °C/1h-annealed sample. As the annealing temperature was further increased to 800 °C, the precipitates within the BCC lamellae disappeared, while the Ni-Al-rich blocky nano-precipitates in the FCC lamellae became coarsened. In the 1000 °C/1h-annealed sample, all precipitates diminished and the elemental distribution within each phase became uniform, despite the significant elemental partitioning between the dual-phase lamellae.

The above results of microstructural evolution and elemental distribution in annealed samples can be used to rationalize the corresponding stress-strain responses in Supplementary Fig. 13. After annealing at low temperatures (600 °C/5h, 660 °C/1h, 700 °C/1h), the nanolamellar structure with an ultrafine interlamellar spacing was still largely retained, but the annealing-induced nano-precipitates could provide extra strengthening and contribute to an extra increase in yield strength. For example, the 600 °C/5h-annealed sample exhibited an exceptional yield strength of $\sigma_{0.2} = 1615 \pm 9$ MPa and ultimate strength of $\sigma_u = 1907 \pm 15$ MPa, along with steady work hardening that resulted in a fairly large uniform elongation above 7%. On the other hand, after annealing at high temperatures (800 °C/1h, 1000 °C/1h), the uniform elongation was increased markedly, while the

yield strength was reduced; the ultimate tensile strength remained at a high level above 1 GPa. This trend is expected to arise from the thermally-induced annihilation of pre-existing dislocations, lamellae coarsening, and even partial recrystallization of the nanolamellar structure.

To summarize, the above heat treatment studies demonstrate that the high metastability of AM materials, as enabled by the extreme processing conditions, open up new opportunities of material design through heat treatment. To harness these opportunities, a clear understanding of the phase transformation pathways in AM materials is essential to develop the heat treatment protocols toward an effective tailoring of microstructure and associated mechanical properties. In this regard, computational modelling through phase-field simulations could be used to understand and predict the microstructure evolution, particularly the evolution of precipitates, under heat treatment. For example, recent phase-field simulations have enabled an understanding of microstructure evolution and phase transformation pathways toward the formation of an “inverted superalloy-like” (i.e., B2 matrix and BCC precipitates) microstructure in a refractory HEA¹¹. For our AM EHEAs, a rich variety of microstructure evolution and phase transformation pathways are possible, including elemental segregation, precipitation, phase ordering, lamellar and precipitate coarsening, and (partial) recrystallization. Some of these transformation pathways have been demonstrated in the aforementioned studies of heat treatment, but more systematic studies through a combined experimental and computational approach are needed in the future.



Supplementary Fig. 16 | Microstructure evolutions of AM AlCoCrFeNi_{2.1} EHEAs after different heat treatments. a, Bright-field TEM images showing the microstructures of as-printed and annealed EHEAs. **b**, STEM-EDX elemental mappings of as-printed and annealed EHEAs.

S5 SUMMARY OF TENSILE PROPERTIES OF ADDITIVELY MANUFACTURED HIGH-PERFORMANCE METAL ALLOYS

Supplementary Table 5 | A summary of tensile properties, including yield strength ($\sigma_{0.2}$), ultimate tensile strength (σ_u), uniform elongation (ϵ_u), and fracture elongation (ϵ_f) of the high-strength metal alloys processed by additive manufacturing. The solid and hollow symbols represent the properties of as-printed and post-annealed samples, respectively.

Composition	AM method	$\sigma_{0.2}$ (MPa)	σ_u (MPa)	ϵ_u (%)	ϵ_f (%)	Symbol
AlCoCrFeNi _{2.1} (this work)	laser powder bed fusion	1333	1640	13.6	15.8	★
		1252	1517	16.6	21.1	
		1615	1907	7.5	7.5	★
		1546	1774	9.4	9.4	
		1453	1694	12.2	12.2	
		1142	1435	17.5	18.0	
		1020	1342	20.2	20.7	
		1003	1286	20.5	23.8	
		823	1200	26.8	31.6	
17-4PH stainless steel ¹²	laser powder bed fusion	945 ± 12	1417 ± 6	11.7 ± 0.8	15.5 ± 1.3	□
		870 ± 25	1358 ± 8	6.5 ± 0.03	13.3 ± 1.5	
		1005 ± 15	1319 ± 2	4.8 ± 0.0	11.1 ± 0.4	
		939 ± 9	1188 ± 6	3.5 ± 0.2	9.0 ± 1.5	
		1352 ± 18	1444 ± 2	2.9 ± 0.1	4.6 ± 0.4	
		1121 ± 9	1172 ± 2	4.8 ± 0.1	9.6 ± 1.7	
		859 ± 11	1017 ± 15	7.7 ± 0.3	16.6 ± 1.2	
15-5PH stainless steel ¹³	laser powder bed fusion	1297 ± 1	1450 ± 2	9	12.53 ± 1.07	△
		1100 ± 39	1467 ± 17	9	14.92 ± 0.32	
4140 steel ¹⁴	laser powder bed fusion	1175 ± 5	1280 ± 10	7	10	●
		1290 ± 10	1325 ± 5	5	7	
H13 tool steel ¹⁵	laser powder bed fusion	1236 ± 178	1717 ± 103	4.1 ± 1.2	4.1 ± 1.2	◆
		835 ± 23	1620 ± 215	4.1 ± 2.3	4.1 ± 2.3	
		1073 ± 72	1965 ± 145	3.7 ± 1.7	3.7 ± 1.7	
HY100 Steel ¹⁶	laser powder bed fusion	1160 ± 15	1200 ± 15	6-7	12 ± 4	◀
Maraging steel ¹⁷	wire arc additive manufacturing	904 ± 78	1118 ± 94	7	11.7 ± 0.8	▼
		1303 ± 29	1410 ± 32	2	8.5 ± 3.3	▽
AF9628 martensitic steel ¹⁸	laser powder bed fusion	1110 ± 10	1400 ± 0	7.2	10.16 ± 1.04	◆
		1110 ± 10	1310 ± 20	5.8	8.95 ± 0.15	
		1110 ± 30	1410 ± 20	6.3	8.38 ± 0.88	
		1040 ± 70	1340 ± 0	4.4	7.40 ± 0.00	
		1080 ± 0	1430 ± 0	7.0	9.79 ± 0.14	
		1080 ± 20	1320 ± 10	6.6	10.89 ± 0.34	
		1100 ± 0	1420 ± 40	6.1	7.71 ± 0.32	
		1080 ± 30	1380 ± 30	6.4	9.51 ± 0.29	
Fe19Ni5Ti ¹⁹	laser directed energy deposition	891	1314	8.8	10.2	◆
		1151	1431	9.0	10.3	
		920	1313	9.0	10.2	

Ti-6Al-4V ²⁰	laser powder bed fusion	1110-1320	1230-1420	2-4	2-14	■
Ti-6Al-2Sn-4Zr-6Mo ²¹	laser powder bed fusion	1052 ± 20 1064 ± 10	1125 ± 18 1146 ± 41	10.9 ± 0.5 10.5 ± 0.5	15.5 ± 0.5 16.4 ± 0.5	◊
Ti-6.5Al-3.5Mo-1.5Zr-0.3Si ²²	laser directed energy deposition	1016 ± 12	1103 ± 12	14 ± 1.1	28.7 ± 1.9	△
Ti-0.4Ni ²³	laser powder bed fusion	849 904 1020 ± 16	924 982 1120 ± 5	11.3 8.7 6.8 ± 0.5	11.5 9.2 8.8 ± 0.5	◆
Ti-6.5Cu ²⁴	laser directed energy deposition	964 ± 31	1073 ± 27	5.5 ± 0.4	7.1	▼
Ti-8.5Cu ²⁴	laser directed energy deposition	1023 ± 29	1180 ± 21	2.1 ± 0.6	4.4	◀
Ti-50Ta ²⁵	laser powder bed fusion	883 ± 20	925 ± 9	8	11.7 ± 1.1	●
Inconel 718 ²⁶⁻²⁸	laser directed energy deposition ²⁶	1084	1333	8	8.4	◊
		1007	1221	16	16	
		949	1194	19	19.9	
	laser powder bed fusion ²⁷	849	1126	19	22.8	◆
		1084	1371	10	10.1	
		1046	1371	12.3	12.3	
	selective electron beam melting ²⁸	925 ± 20	1138 ± 24	15	15.7 ± 4.3	◆
		894 ± 24	1061 ± 83	11.5	11.5 ± 6.9	
		1061 ± 16 1035 ± 17	1266 ± 44 1240 ± 19	15 17	21.1 ± 1.1 21.8 ± 2.4	
CoCrW ²⁹	laser powder bed fusion	850 ± 23	1158 ± 21	9.8 ± 0.1	9.8 ± 0.1	◀
Co _{1.5} CrFeNi _{1.5} Ti _{0.5} Mo _{0.1} ³⁰	selective electron beam melting	867 ± 2	1323 ± 5	17 ± 2	17 ± 2	□
Ni _{2.1} CoCrFeNb _{0.2} ³¹	laser directed energy deposition	896 873	1127 1045	15.3 12.6	16.8 13.9	▼
Al _{0.2} Co _{1.5} CrFeNi _{1.5} Ti _{0.3} ³²	laser powder bed fusion	1235	1550	10.69	10.69	◻
Zr _{39.6} Ti _{33.9} Nb _{7.6} Cu _{6.4} Be _{12.5} ³³	laser directed energy deposition	1131	1178	2.4	12	■
		1430	1487	2.4	2.7	
		1317	1459	4.1	13	

S6 DP-FE MODELLING AND SIMULATIONS

1. Micromechanics model

Based on neutron diffraction data (i.e., diffraction elastic constants), we developed a micromechanics model to inversely determine the anisotropic elastic constants of co-deforming FCC and BCC phases in our AM EHEAs, which enabled us to analyse stress partition in the two phases. First, we derived an analytic micromechanics solution of diffraction elastic constants for co-deforming FCC and BCC phases by assuming their elastic properties are known. Then, we used the micromechanics solution to solve an optimization problem for determining the anisotropic elastic constants of both FCC and BCC phases in our AM EHEAs.

To derive the micromechanics solution, we considered a spherical FCC grain α (representing an FCC colony) in a $\{hkl\}$ crystallographic family embedded in an elastically isotropic matrix comprising a mixture of FCC and BCC phases. Suppose the matrix has the effective bulk modulus $\bar{K}$ and shear modulus $\bar{\mu}$. Based on the previous analytic solution³⁴, we used the constrained compliance tensor $\mathbf{U}_{(\alpha)}^{\text{FCC}}$ to relate the applied average stress tensor $\bar{\boldsymbol{\sigma}}$ and local strain tensor $\boldsymbol{\varepsilon}_{(\alpha)}^{\text{FCC}}$ in the FCC grain α according to

$$\boldsymbol{\varepsilon}_{(\alpha)}^{\text{FCC}} = \mathbf{U}_{(\alpha)}^{\text{FCC}} \bar{\boldsymbol{\sigma}}$$

Here the fourth-order tensor $\mathbf{U}_{(\alpha)}^{\text{FCC}}$ can be expressed in a compact symbolic form

$$\mathbf{U}_{(\alpha)}^{\text{FCC}} = \left(\frac{1}{3[\bar{K} + \bar{\gamma}(3\eta_1^{\text{FCC}} - 3\bar{K})]}, \frac{1}{2[\bar{\mu} + \bar{\delta}(2\eta_2^{\text{FCC}} - 2\bar{\mu})]}, \frac{1}{2[\bar{\mu} + \bar{\delta}(2\eta_3^{\text{FCC}} - 2\bar{\mu})]} \right)$$

where $\bar{\gamma} = \frac{\bar{K}}{3\bar{K} + 4\bar{\mu}}$ and $\bar{\delta} = \frac{3\bar{K} + 6\bar{\mu}}{15\bar{K} + 20\bar{\mu}}$ are the components of the Eshelby tensor for a spherical

inclusion³⁴; $3\eta_1^{\text{FCC}} = C_{11}^{\text{FCC}} + 2C_{12}^{\text{FCC}}$, $2\eta_2^{\text{FCC}} = C_{11}^{\text{FCC}} - C_{12}^{\text{FCC}}$, $2\eta_3^{\text{FCC}} = 2C_{11}^{\text{FCC}}$ are the combinations of FCC elastic constants. The above expression of $\mathbf{U}_{(\alpha)}^{\text{FCC}}$ can be rewritten as

$$\mathbf{U}_{(\alpha)}^{\text{FCC}} = (3a^{\text{FCC}}, 2b^{\text{FCC}}, 2c^{\text{FCC}})$$

where a^{FCC} , b^{FCC} and c^{FCC} are introduced to represent the corresponding terms in $\mathbf{U}_{(\alpha)}^{\text{FCC}}$ and they are dependent on both the FCC elastic constants C_{11}^{FCC} , C_{12}^{FCC} , C_{44}^{FCC} and the effective elastic moduli of the isotropic matrix $\bar{K}$, $\bar{\mu}$. From $\mathbf{U}_{(\alpha)}^{\text{FCC}}$, we obtained the analytic solution of diffraction elastic constants of grain α along the loading direction (LD)

$$\frac{1}{E_{\text{LD}}^{\text{FCC}, hkl}} = \frac{\langle \mathbf{n} \cdot \boldsymbol{\varepsilon}^{\text{FCC}} \cdot \mathbf{n} \rangle_{hkl}}{\mathbf{n} \cdot \bar{\boldsymbol{\sigma}} \cdot \mathbf{n}} = \frac{3a^{\text{FCC}} + 4b^{\text{FCC}}}{3} - 4(b^{\text{FCC}} - c^{\text{FCC}})\Gamma$$

where $\mathbf{n}$ is the unit vector along the loading direction, $\langle \bullet \rangle_{hkl}$ represents the average over grains with their $[hkl]$ directions aligned with $\mathbf{n}$, and $\Gamma = (h^2k^2 + l^2k^2 + h^2l^2)/(h^2 + k^2 + l^2)^2$ is the orientation index parameter. Following a similar procedure, we obtained the analytic solution of diffraction elastic constants of the BCC phase along LD,

$$\frac{1}{E_{\text{LD}}^{\text{BCC}, hkl}} = \frac{\langle \mathbf{n} \cdot \boldsymbol{\varepsilon}^{\text{BCC}} \cdot \mathbf{n} \rangle_{hkl}}{\mathbf{n} \cdot \bar{\boldsymbol{\sigma}} \cdot \mathbf{n}} = \frac{3a^{\text{BCC}} + 4b^{\text{BCC}}}{3} - 4(b^{\text{BCC}} - c^{\text{BCC}})\Gamma$$

where a^{BCC} , b^{BCC} and c^{BCC} are introduced to represent the constrained compliance tensor $\mathbf{U}_{(\alpha)}^{\text{BCC}}$ and they are dependent on the elastic constants of the BCC phase C_{11}^{BCC} , C_{12}^{BCC} , C_{44}^{BCC} and the effective elastic moduli of the isotropic matrix $\bar{K}$, $\bar{\mu}$. We also obtained the similar analytic solution of diffraction elastic constants along the transverse direction (TD). We note that the elastic constants

of $C_{11}^{\text{FCC}}, C_{12}^{\text{FCC}}, C_{44}^{\text{FCC}}, C_{11}^{\text{BCC}}, C_{12}^{\text{BCC}}, C_{44}^{\text{BCC}}, \bar{K}, \bar{\mu}$ completely determine the diffraction elastic constants of FCC and BCC phases, but they are not all independent of each other. Namely, we invoked the self-consistent method to evaluate the effective moduli $\bar{K}, \bar{\mu}$ for our dual-phase EHEAs. Hence, $\bar{K}, \bar{\mu}$ can be determined from the elastic constants of $C_{11}^{\text{FCC}}, C_{12}^{\text{FCC}}, C_{44}^{\text{FCC}}, C_{11}^{\text{BCC}}, C_{12}^{\text{BCC}}, C_{44}^{\text{BCC}}$ and the volume fractions of FCC and BCC phases. Specifically, the average strain for a dual-phase EHEA over all grain orientations is given by

$$\begin{aligned}\bar{\boldsymbol{\varepsilon}} &= \phi^{\text{FCC}} \langle \boldsymbol{\varepsilon}^{\text{FCC}} \rangle_{\text{all}} + \phi^{\text{BCC}} \langle \boldsymbol{\varepsilon}^{\text{BCC}} \rangle_{\text{all}} \\ &= \left[\phi^{\text{FCC}} \langle \mathbf{U}^{\text{FCC}} \rangle_{\text{all}} + \phi^{\text{BCC}} \langle \mathbf{U}^{\text{BCC}} \rangle_{\text{all}} \right] \bar{\boldsymbol{\sigma}} \\ &= \left(3\phi^{\text{FCC}} a^{\text{FCC}} + 3\phi^{\text{BCC}} a^{\text{BCC}}, \frac{4\phi^{\text{FCC}} b^{\text{FCC}} + 4\phi^{\text{BCC}} b^{\text{BCC}} + 6\phi^{\text{FCC}} c^{\text{FCC}} + 6\phi^{\text{BCC}} c^{\text{BCC}}}{5} \right) \bar{\boldsymbol{\sigma}} \\ &= \left(3\bar{a}, \frac{4\bar{b} + 6\bar{c}}{5} \right) \bar{\boldsymbol{\sigma}}\end{aligned}$$

where ϕ^{FCC} and ϕ^{BCC} are the volume fractions of FCC and BCC phases, respectively, and $\bar{a}, \bar{b}, \bar{c}$ are the phase average of previously defined a, b, c quantities. Recall that the average compliance tensor $\bar{\mathbf{M}}$ is given by

$$\bar{\boldsymbol{\varepsilon}} = \bar{\mathbf{M}} \bar{\boldsymbol{\sigma}} = \left(\frac{1}{3\bar{K}}, \frac{1}{2\bar{\mu}} \right) \bar{\boldsymbol{\sigma}}$$

As such, the above result expresses $\bar{K}, \bar{\mu}$ in terms of $C_{11}^{\text{FCC}}, C_{12}^{\text{FCC}}, C_{44}^{\text{FCC}}, C_{11}^{\text{BCC}}, C_{12}^{\text{BCC}}, C_{44}^{\text{BCC}}$. This self-consistent result is enforced by comparing the predicted values of $\bar{K}, \bar{\mu}$ with the corresponding experimental values.

Next, we used the above micromechanics solution to solve an optimization problem for determining the anisotropic elastic constants of FCC and BCC phases in our AM EHEAs. From the experimentally measured diffraction elastic constants and elastic moduli of AM EHEAs, we employed the gradient descent algorithm to minimize the error between predicted and measured values. We started with a random initial guess of $\mathbf{x} = [C_{11}^{\text{FCC}}, C_{12}^{\text{FCC}}, C_{44}^{\text{FCC}}, C_{11}^{\text{BCC}}, C_{12}^{\text{BCC}}, C_{44}^{\text{BCC}}, \bar{K}, \bar{\mu}]$. Then we computed the target elastic properties arranged in a linear array of $\mathbf{y}$, i.e.,

$$\begin{aligned}\mathbf{y} &= \mathbf{f}(\mathbf{x}) \\ &= [E_{\text{LD}}^{\text{FCC},200}, E_{\text{LD}}^{\text{FCC},220}, E_{\text{LD}}^{\text{FCC},111}, E_{\text{LD}}^{\text{FCC},311}, E_{\text{LD}}^{\text{FCC},331}, \\ &\quad E_{\text{TD}}^{\text{FCC},200}, E_{\text{TD}}^{\text{FCC},220}, E_{\text{TD}}^{\text{FCC},111}, E_{\text{TD}}^{\text{FCC},311}, E_{\text{TD}}^{\text{FCC},331}, \\ &\quad E_{\text{LD}}^{\text{BCC},200}, E_{\text{LD}}^{\text{BCC},110}, E_{\text{LD}}^{\text{BCC},211}, E_{\text{LD}}^{\text{BCC},321}, \\ &\quad E_{\text{TD}}^{\text{BCC},200}, E_{\text{TD}}^{\text{BCC},110}, E_{\text{TD}}^{\text{BCC},211}, E_{\text{TD}}^{\text{BCC},321}, \\ &\quad 3\bar{K} - \frac{1}{3\bar{a}}, 2\bar{\mu} - \frac{5}{4\bar{b} + 6\bar{c}}, \frac{9\bar{K}\bar{\mu}}{3\bar{K} + \bar{\mu}}]\end{aligned}$$

where the first 18 components in $\mathbf{y}$ are the diffraction elastic constants for different grain families along LD and TD, and the last 3 components serve to enforce the self-consistency for the dual-phase polycrystal, as discussed earlier. The target values of $\mathbf{y}$ are given as $\mathbf{y}_0$ taken from experimental measurements, i.e.,

$$\mathbf{y}_0 = [136.53, 208.64, 238.72, 175.14, 222.78, \\ 399.34, 579.90, 784.18, 525.59, 646.36, \\ 107.75, 197.31, 192.44, 192.13, \\ 274.29, 542.89, 511.56, 491.94, \\ 0, 0, 171] \text{ GPa}$$

To find the optimal solution for the elastic constants of FCC and BCC phases, the loss function L is defined as

$$L = (\mathbf{y} - \mathbf{y}_0)\mathbf{g}(\mathbf{y} - \mathbf{y}_0)^T$$

where the weight matrix is $\mathbf{g} = \text{diag}(4, 4, 4, 4, 4, 1, 1, 1, 1, 1, 4, 4, 4, 4, 1, 1, 1, 1, 3, 5, 3)$. This weight matrix was introduced to balance the relative importance of different diffraction elastic constants because their values along TD are usually two times higher than those along LD. The gradient descent algorithm was used to minimize the loss function L by updating the components in $\mathbf{x}_i$ at the i -th iteration step along the gradient direction,

$$\mathbf{x}_{i+1} = \mathbf{x}_i - \lambda \frac{\partial L}{\partial \mathbf{x}}$$

The gradient $\partial L / \partial \mathbf{x}$ was calculated numerically by small perturbations near $\mathbf{x}_i$ and the iteration step size λ is 1×10^{-3} . The iteration process of optimization was terminated when the norm of $\partial L / \partial \mathbf{x}$ is less than 1×10^{-3} . Three different sets of initial values of $\mathbf{x}_0$ were tested and they all converged to the same result. The final optimized elastic constants of FCC and BCC phases for our AM EHEAs are listed in Supplementary Table 6.

Supplementary Table 6 | Elastic constants of FCC and BCC phases in the AlCoCrFeNi_{2.1} EHEA determined from the micromechanics model (in GPa).

C_{11}^{FCC}	C_{12}^{FCC}	C_{44}^{FCC}	C_{11}^{BCC}	C_{12}^{BCC}	C_{44}^{BCC}	$\bar{K}$	$\bar{\mu}$
258.72	177.52	124.82	282.44	259.75	131.90	225.03	63.37

2. DP-CPFE model

A DP-CPFE model was developed to evaluate both linear and non-linear lattice strain responses as well as to perform stress partition analysis in the dual-phase EHEA. We used the rate-dependent, finite strain, crystal plasticity constitutive formulation for both FCC and BCC phases. The deformation gradient $\mathbf{F}$ is decomposed into $\mathbf{F} = \mathbf{F}^e \mathbf{F}^p$, where $\mathbf{F}^e$ is the elastic deformation gradient and $\mathbf{F}^p$ is the plastic deformation gradient. The elastic Green strain tensor can be defined as $\mathbf{E}^e = 1/2(\mathbf{F}^{eT} \mathbf{F}^e - \mathbf{I})$, where $\mathbf{I}$ is second-order identity tensor. The second Piola-Kirchhoff stress $\mathbf{T}^*$ in the intermediate configuration can be determined as $\mathbf{T}^* = \mathbf{C}:\mathbf{E}^e$, and $\mathbf{C}$ is the forth-order single-crystal elastic tensor. The rate of plastic deformation gradient is given by $\dot{\mathbf{F}}^p = \mathbf{L}^p \mathbf{F}^p$. Here, the plastic velocity gradient is expressed as $\mathbf{L}^p = \sum_{i=1}^{12} \dot{\gamma}_i^p \mathbf{m}_i \otimes \mathbf{n}_i$, where $\dot{\gamma}_i^p$ is the plastic shearing rate on i -th slip system, $\mathbf{m}_i$ and $\mathbf{n}_i$ are the slip direction and slip plane normal, respectively. We considered 12 $\{111\}\langle 110 \rangle$ slip systems in the FCC phase and 12 $\{110\}\langle 111 \rangle$ in the BCC phase. The plastic shearing rate $\dot{\gamma}_i^p$ is given by a power-law relation, $\dot{\gamma}_i^p = \dot{\gamma}_0^p |\tau_i / s_i|^{1/m} \text{sgn}(\tau_i)$, where $\dot{\gamma}_0^p$ is the reference shearing rate, $\tau_i = \mathbf{m}_i \cdot \mathbf{T} \cdot \mathbf{n}_i$ is the resolved shear stress, m is the rate sensitivity and s_i is the slip resistance on the i -th slip system. The slip resistance on each slip system starts

with an identical initial value s_0 and is hardened according to $\dot{s}_i = \sum_j h_{ij} |\dot{\gamma}_j^p|$; the hardening matrix is expressed as $h_{ij} = q_{ij} h_0 (1 - s_j / s_{\text{sat}})^a$, where the diagonal components of q_{ij} are 1.0 and the off-diagonal components are 1.4. The parameters used in the above constitutive model are listed in Supplementary Table 7. The large h_0 values are mainly associated with the hardening rate due to dislocation interactions between different slip systems in the confined nanolamellae and they are not necessarily controlled by the higher initial density of dislocations.

Supplementary Table 7 | Parameters used in the DP-CPFE model

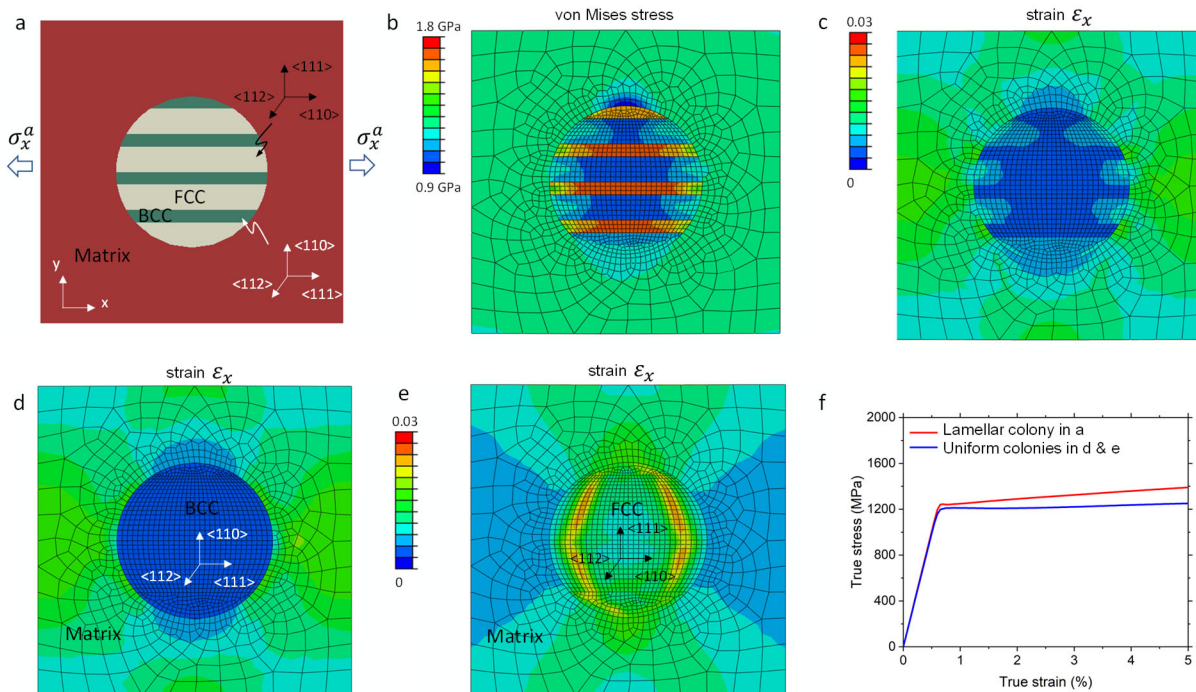
Phase	$\dot{\gamma}_0^p$ (/s)	m	s_0 (MPa)	h_0 (MPa)	s_{sat} (MPa)	a
FCC	0.001	0.02	550	1000	600	0.7
BCC	0.001	0.02	475	4000	750	0.7

The above constitutive model was implemented in the general finite element package ABAQUS/Explicit by writing a user-defined material subroutine VUMAT. In finite element simulations, we constructed a polycrystalline cube comprising 8000 elements, where each element is treated as a single-crystal grain (representing an FCC or BCC colony). The crystallographic orientation of each grain was randomly assigned to generate a texture-free sample. Each grain was also randomly assigned as an FCC or BCC phase, such that the corresponding volume fraction of each phase is consistent with experimental values. The polycrystal structure was subjected to uniaxial loading with a strain rate of 0.001 /s. The lattice strain response was extracted from the elastic strain tensor in each grain/colony by writing a Python script through the ABAQUS-Python interface.

In the above DP-CPFE model, the FCC and BCC phases are treated as separate colonies/grains with different orientations, such that they are not explicitly coupled to account for the constraining effects of alternating FCC and BCC lamellar structures. To evaluate the constraining effects on our DP-CPFE results, we performed additional DP-CPFE simulations. As shown in Supplementary Fig. 17a, we constructed a model consisting of an FCC/BCC lamellar colony/grain embedded in a homogenized poly-colony/grain matrix under an applied tensile stress σ_x^a . The volume fractions of FCC and BCC lamellae match those in our AM EHEA. In this model, the experimentally observed Kurdjumov-Sachs (K-S) orientation relationship of $\{111\}_{\text{FCC}} \parallel \{110\}_{\text{BCC}}$ and $\langle 110 \rangle_{\text{FCC}} \parallel \langle 111 \rangle_{\text{BCC}}$ is enforced between the neighbouring FCC and BCC lamellae. That is, given a specific orientation of FCC lamellae in the global coordinate system, the corresponding orientation of BCC lamellae is determined by the K-S relationship. In addition, this lamellar structure imposes mutual geometrical constraints between neighbouring FCC/BCC lamellae during their progressive elasto-plastic deformation under applied mechanical loading. The same set of CP constitutive relations and parameters (Supplementary Table 7) are used for the FCC and BCC lamellae. In addition, a J_2 plasticity model is used to represent the homogenized stress-strain behaviour of the poly-colony/grain matrix. The corresponding J_2 plasticity parameters are determined by fitting to the experimentally measured stress-strain curve of the AM EHEA. Some representative DP-CPFE results of the model in Supplementary Fig. 17a are presented in Supplementary Fig. 17b and c, which respectively show the non-uniform distribution of von Mises stresses and tensile strains ε_x in FCC and BCC lamellae.

Similar to Supplementary Fig. 17a, we constructed a model consisting of a single BCC colony/grain embedded in a homogenized poly-colony/grain matrix under an applied tensile stress σ_x^a , as shown in Supplementary Fig. 17d. This BCC colony/grain has the same orientation and CP parameters as the BCC lamellae in the FCC/BCC lamellar colony/grain in Supplementary Fig. 17a. The aforementioned J_2 plasticity model is used to represent the stress-strain behaviour of the homogenized poly-colony/grain matrix. The DP-CPFE simulation result of this model is presented in Supplementary Fig. 17d, which shows the uniform distribution of tensile strain ε_x in the BCC colony/grain. In addition, we constructed a similar model consisting of a single FCC colony/grain embedded in a homogenized poly-colony/grain matrix under an applied tensile stress σ_x^a . The DP-CPFE simulation result of this model is presented in Supplementary Fig. 17e, which shows the relatively non-uniform distribution of tensile strain ε_x in the FCC colony/grain compared to its BCC counterpart.

More importantly, Supplementary Fig. 17f shows the effective/average elastic-plastic curve (red line) of the FCC/BCC lamellar colony/grain in Supplementary Fig. 17a. Calculations of the effective/average stress in the FCC/BCC lamellar colony/grain require the post-processing of Abaqus DP-CPFE output by writing a Python script. In addition, Supplementary Fig. 17f shows the volume-fraction-averaged elastic-plastic curve (blue line) of the FCC and BCC colony/grain from the model in Supplementary Fig. 17d and e, which is reasonably consistent with that of the FCC/BCC lamellar colony/grain from the model of Supplementary Fig. 17a. Small quantitative differences between the two curves reflect the constraining effects of alternating FCC and BCC lamellar structures. Hence, this comparison study demonstrates that our DP-CPFE model can closely capture the effective stress-strain behaviour in the respective FCC and BCC phases in nanolamellar eutectic colonies. They are the main deformation characteristics that we focus on in this work. The spatially non-uniform distributions of stresses and strains in each phase are more sensitive to the constraining effects between alternating FCC and BCC lamellae, and their impact on the DP-CPFE modelling warrants further study in the future.

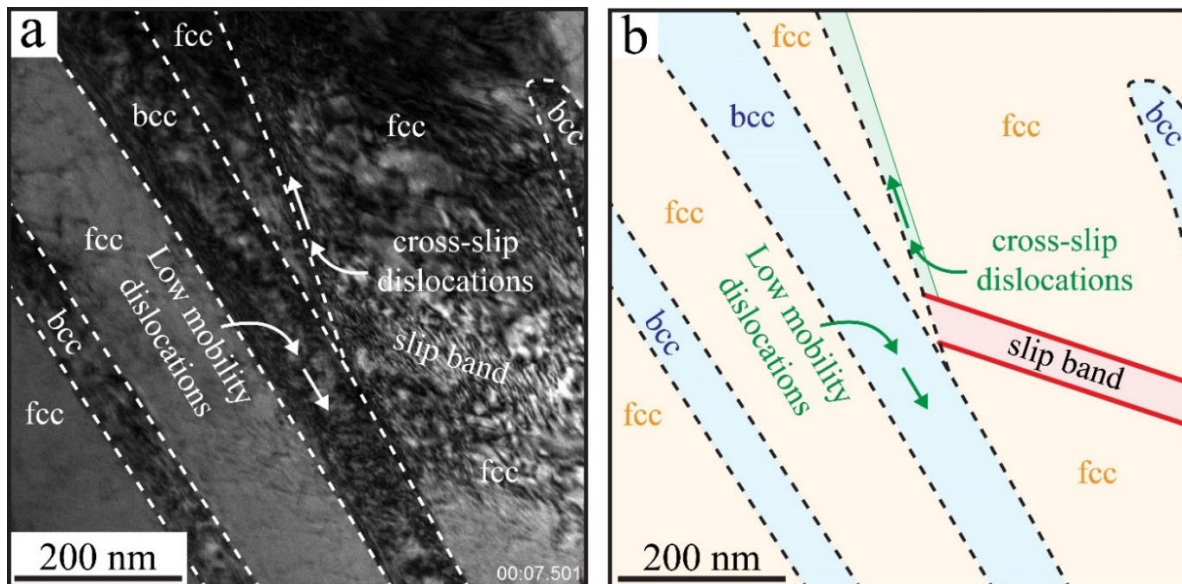


Supplementary Fig. 17 | DP-CPFE comparison between the average stress-strain response of a colony/grain containing an FCC/BCC lamellar structure and the average stress-strain response of an FCC and a BCC colony/grain, as embedded in the same homogenized poly-colony/grain matrix under an applied tensile stress σ_x^a . **a**, Schematic of a colony/grain embedded in an effective poly-colony/grain matrix. This representative colony/grain contains lamellar microstructures with alternating FCC and BCC phases, where the K-S orientation relationship is enforced. **b**, DP-CPFE contour of the von Mises stress in an FCC/BCC lamellar colony/grain. **c**, DP-CPFE contour of the normal strain ϵ_x (along the loading direction) in an FCC/BCC lamellar colony/grain. **d**, Same as **c**, except for a uniform BCC colony/grain. **e**, Same as **c**, except for a uniform FCC colony/grain. **d** and **e** share the same contour legend. **f**, Comparison of the average stress in the lamellar colony/grain vs. applied strain in **a** against the volume-fraction-weighted average stress of the FCC colony/grain in **d** and BCC colony/grain in **e** vs. applied strain.

S7 DESCRIPTION OF *IN SITU* TEM VIDEO

Supplementary Video 1: *In situ* TEM showing dislocation movements in BCC and FCC nanolamellae under tensile loading

In situ TEM mechanical tests were conducted in an FEI 80-300 Titan in TEM mode. 11 mm long, 3 mm wide, and 100 μm thick dog-bone shaped tensile specimens were cut by wire EDM from as-built parts. They were then electropolished following the same recipe described in the Methods section. The specimens were then loaded in tension at a maximum speed of 1 $\mu\text{m/s}$ using a GATAN straining holder, model 654. From *in situ* TEM straining experiment, we observed that the dislocation mobility in the FCC lamellae is much higher than that in the BCC lamellae (Supplementary Video 1 and Fig. 18). Interestingly, dislocations in the FCC lamellae were observed to cross-slip and move along the interface. Note that dislocations in the slip bands keep multiplying but their density seems to remain stable, indicating active dislocation annihilation at the interface. In contrast, dislocations in the BCC lamellae required the elevated local stresses to nucleate and move, e.g., through dislocation pile-ups in the adjacent FCC lamellae. This is shown at the end of Supplementary Video 1 when dislocations in the BCC lamellae were activated near the tip of the slip band in the FCC lamellae. Moreover, motion of the dislocations in the BCC lamellae was more sluggish. They tended to stop ahead of the lamellar interface when they approached the limit of the zone affected by the pile-ups in the adjacent FCC lamellae. These observations shed light on the mechanistic origin leading to the differences in dislocation densities between the two phases. Namely, unlike in FCC, dislocations in BCC lamellae can nucleate but escape the lamellae more rarely. The lower mobility of the dislocations in the BCC nanolamellae is understandable considering the generally higher Peierls barriers to dislocation glide in BCC metals. In addition, the nanoscale chemical modulation in the BCC lamellae is expected to further enhance the lattice distortion arising from atomic size misfits. As such, the ruggedness of the energy landscape controlling dislocation mobility becomes heightened, making dislocation motion in the BCC lamellae more sluggish and thereby enhancing dislocation storage.



Supplementary Fig. 18 | Deforming nanolamellar structure (a) and schematic illustration (b) in an *in situ* TEM straining experiment (Supplementary Video 1).

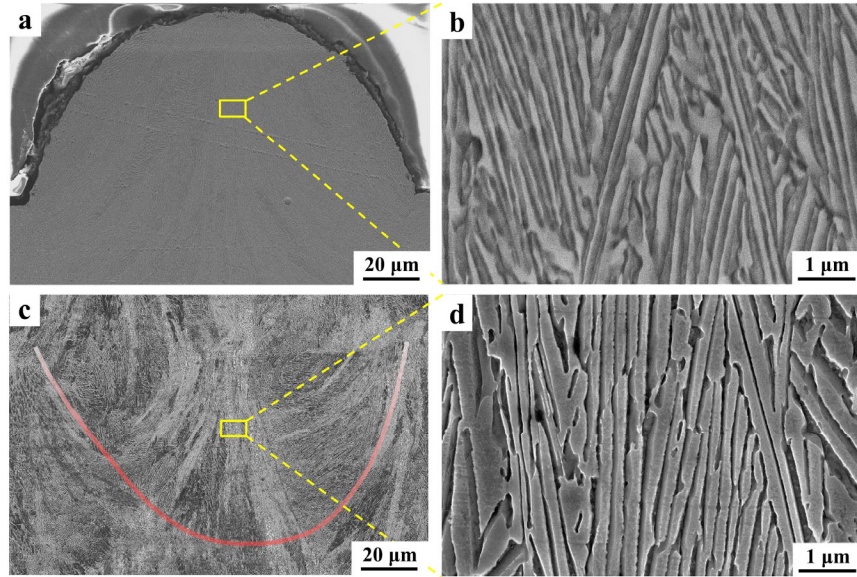
S8 *IN SITU* THERMAL CYCLING EFFECT

In situ thermal cycling, also called intrinsic heat treatment, is a unique characteristic inherited to the additive manufacturing process. *In situ* thermal cycling during repeated layer build can produce heterogeneity that critically affects the microstructure and associated properties of AM alloys. It is noteworthy that such heterogeneity can be exploited to design novel metal alloys with enhanced mechanical properties as demonstrated in several pioneering works^{19,24}.

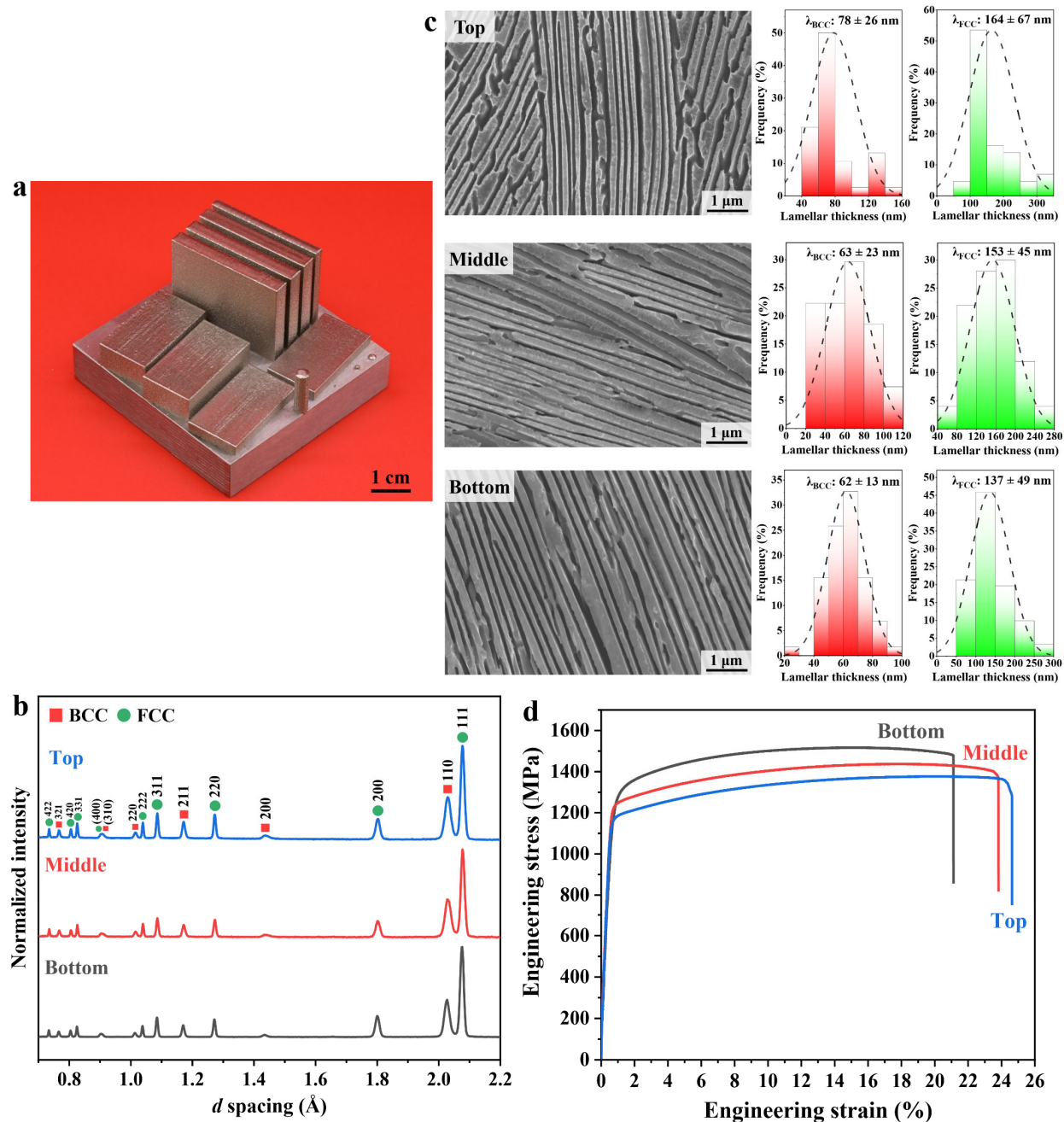
In order to characterize and understand the effects of intrinsic heat treatment on microstructure, we performed a series of studies as follows. Firstly, we printed a single-track bead (on the same AlCoCrFeNi_{2.1} EHEA substrate) as a comparative case without repeated heat transfer. In Supplementary Fig. 19, we compare the microstructures of the single-track bead and the multi-layer bulk specimen taken from the middle of a 2-mm-height rectangular plate produced by the same laser processing strategy. Both the single-track bead and bulk specimen feature the eutectic nanolamellar structures with an average interlamellar spacing of 208 nm and 215 nm, respectively. This comparison indicates a negligible effect of intrinsic heat treatment on the formation of nanolamellar structure in the AM EHEA.

Secondly, we printed much taller thin walls with a dimension of 30 mm (length) $\times$ 3.5 mm (width) $\times$ 25 mm (height) (Supplementary Fig. 20a). We carefully analysed the phase constituents and microstructures in the top, middle, and bottom sections of these tall walls. For example, neutron diffraction experiments were conducted for samples at different build heights, showing no appreciable changes of crystal structure or phase fraction along the build direction (Supplementary Fig. 20b). These tall thin-walls exhibit the characteristic nanolamellar structure throughout the entire build height (Supplementary Fig. 20c). With an increase of build height, the interlamellar spacing becomes slightly larger, which can be attributed to the increased heat accumulation and decreased cooling rate with increasing height. Such microstructure changes, however, result in only small changes in the overall mechanical properties (Supplementary Fig. 20d). These different combinations of yield strength and ductility at different build heights fall into the mechanical property space of AM EHEAs highlighted in Fig. 2b of the main text. In addition, the site-specific cooling rate and its dependence on build height have been studied in literature^{3,35,36}. These studies show that the large heat sink of the printing substrate often leads to a locally enhanced cooling rate at the bottom section of a build. For the top section of the build, heat is mainly conducted through the previously deposited layers. As such, the accumulation of heat during repeated melting, along with longer heat dissipation paths, can result in a gradually reduced cooling rate with increasing build height. Nevertheless, such differences in the cooling rate are usually small. For example, Pauly et al.³ reported a similar trend in a 6.5-mm-tall selectively laser melted Al-33Cu (wt.%) eutectic alloy. The cooling rate at the bottom was slightly higher than that at the top, but still comparable on the same order of $\sim 10^5$ K/s.

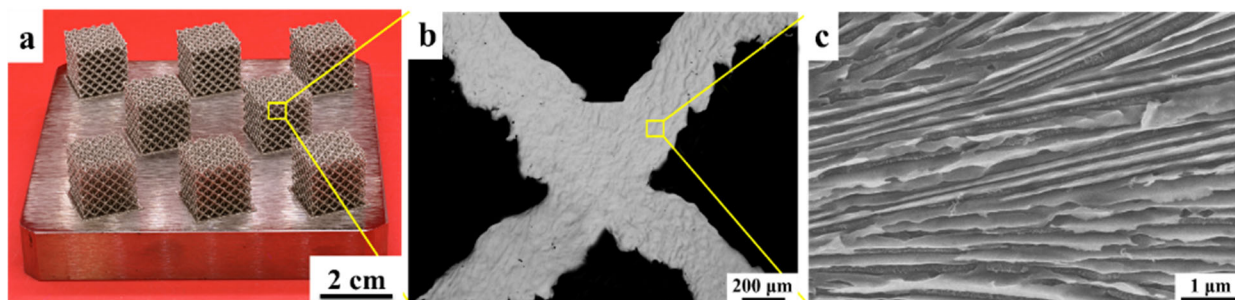
Finally, we examined the microstructure in a more complex structure than a simple block. As shown in Supplementary Fig. 21, we printed a batch of octet micro-lattices with a strut diameter of approximately 300 μ m; the characteristic nanolamellae with an interlamellar spacing of 182 nm form in these micro-lattices, suggesting the excellent printability and the reproducible control of nanolamellar structures for the AlCoCrFeNi_{2.1} EHEA.



Supplementary Fig. 19 | Secondary electron images showing the microstructure in an as-printed single-track bead (a, b) and a multi-layer bulk sample (c, d). b and d show the magnified microstructures within the yellow boxed regions in a and c, respectively. The melt pool boundary in c is outlined by the red solid line.



Supplementary Fig. 20 | Characterization of microstructures and mechanical properties along the build direction of as-printed AlCoCrFeNi_{2.1} thin-wall structures. a, Printed thin-wall structures. **b**, Neutron diffraction patterns of the top, middle and bottom sections of the as-printed thin wall. **c**, Secondary electron images showing similar nanolamellar microstructures and BCC and FCC lamellar thickness histograms for the top, middle and bottom sections of the as-printed thin wall. **d**, Tensile stress-strain curves of the top, middle and bottom sections of the as-printed thin wall.



Supplementary Fig. 21 | Microstructure of as-printed $\text{AlCoCrFeNi}_{2.1}$ octet lattices. **a**, Printed octet lattices. **b**, Optical micrograph showing the cross-section of the octet lattice with the truss diameter ~ 300 μm . **c**, Secondary electron image showing the nanolamellar structures of the octet lattice.

Supplementary References

1. Thomas, M., Baxter, G. J. & Todd, I. Normalised model-based processing diagrams for additive layer manufacture of engineering alloys. *Acta Materialia* **108**, 26–35 (2016).
2. Jackson, K. A. & Hunt, J. D. Lamellar and rod eutectic growth. in *Dynamics of Curved Fronts* 363–376 (Elsevier, 1988).
3. Pauly, S., Wang, P., Kühn, U. & Kosiba, K. Experimental determination of cooling rates in selectively laser-melted eutectic Al-33Cu. *Additive Manufacturing* **22**, 753–757 (2018).
4. Lu, Y. *et al.* Directly cast bulk eutectic and near-eutectic high entropy alloys with balanced strength and ductility in a wide temperature range. *Acta Materialia* **124**, 143–150 (2017).
5. Withers, P. J. & Bhadeshia, H. Residual stress. Part 1—measurement techniques. *Materials science and Technology* **17**, 355–365 (2001).
6. Withers, P. J. & Bhadeshia, H. Residual stress. Part 2—Nature and origins. *Materials science and technology* **17**, 366–375 (2001).
7. Collins, P. C., Brice, D. A., Samimi, P., Ghamarian, I. & Fraser, H. L. Microstructural control of additively manufactured metallic materials. *Annual Review of Materials Research* **46**, 63–91 (2016).
8. Bartlett, J. L. & Li, X. An overview of residual stresses in metal powder bed fusion. *Additive Manufacturing* **27**, 131–149 (2019).
9. Chen, W. *et al.* Microscale residual stresses in additively manufactured stainless steel. *Nature Communications* **10**, 4338 (2019).
10. Williamson, G. K. & Hall, W. H. X-ray line broadening from filed aluminium and wolfram. *Acta Metallurgica* **1**, 22–31 (1953).
11. Kadirvel, K., Kloenne, Z., Jensen, J. K., Fraser, H. & Wang, Y. Phase-field modelling of transformation pathways and microstructural evolution in multi-principal element alloys. *Applied Physics Letters* **119**, 171905 (2021).
12. LeBrun, T., Nakamoto, T., Horikawa, K. & Kobayashi, H. Effect of retained austenite on subsequent thermal processing and resultant mechanical properties of selective laser melted 17–4 PH stainless steel. *Materials & Design* **81**, 44–53 (2015).
13. Rafi, H. K., Starr, T. L. & Stucker, B. E. A comparison of the tensile, fatigue, and fracture behavior of Ti–6Al–4V and 15-5 PH stainless steel parts made by selective laser melting. *The International Journal of Advanced Manufacturing Technology* **69**, 1299–1309 (2013).
14. Damon, J. *et al.* Process development and impact of intrinsic heat treatment on the mechanical performance of selective laser melted AISI 4140. *Additive Manufacturing* **28**, 275–284 (2019).
15. Mertens, R. *et al.* Influence of powder bed preheating on microstructure and mechanical properties of H13 tool steel SLM parts. *Physics Procedia* **83**, 882–890 (2016).
16. Dilip, J. J. S., Ram, G. J., Starr, T. L. & Stucker, B. Selective laser melting of HY100 steel: process parameters, microstructure and mechanical properties. *Additive Manufacturing* **13**, 49–60 (2017).
17. Xu, X. *et al.* Microstructural evolution and mechanical properties of maraging steel produced by wire+ arc additive manufacture process. *Materials Characterization* **143**, 152–162 (2018).
18. Seede, R. *et al.* An ultra-high strength martensitic steel fabricated using selective laser melting additive manufacturing: Densification, microstructure, and mechanical properties. *Acta Materialia* **186**, 199–214 (2020).
19. Kürnsteiner, P. *et al.* High-strength Damascus steel by additive manufacturing. *Nature* **582**, 515–519 (2020).

20. Voisin, T. *et al.* Defects-dictated tensile properties of selective laser melted Ti-6Al-4V. *Materials & Design* **158**, 113–126 (2018).
21. Carrozza, A., Aversa, A., Fino, P. & Lombardi, M. A study on the Microstructure and Mechanical Properties of the Ti-6Al-2Sn-4Zr-6Mo Alloy Produced via Laser Powder Bed Fusion. *Journal of Alloys and Compounds* 159329 (2021).
22. Zhu, Y., Li, J., Tian, X., Wang, H. & Liu, D. Microstructure and mechanical properties of hybrid fabricated Ti–6.5 Al–3.5 Mo–1.5 Zr–0.3 Si titanium alloy by laser additive manufacturing. *Materials Science and Engineering: A* **607**, 427–434 (2014).
23. Xiong, Z., Pang, X., Liu, S., Li, Z. & Misra, R. D. K. Hierarchical refinement of nickel-microalloyed titanium during additive manufacturing. *Scripta Materialia* **195**, 113727 (2021).
24. Zhang, D. *et al.* Additive manufacturing of ultrafine-grained high-strength titanium alloys. *Nature* **576**, 91–95 (2019).
25. Sing, S. L., Yeong, W. Y. & Wiria, F. E. Selective laser melting of titanium alloy with 50 wt% tantalum: Microstructure and mechanical properties. *Journal of Alloys and Compounds* **660**, 461–470 (2016).
26. Qi, H., Azer, M. & Ritter, A. Studies of standard heat treatment effects on microstructure and mechanical properties of laser net shape manufactured Inconel 718. *Metallurgical and Materials Transactions A* **40**, 2410–2422 (2009).
27. Zhang, D., Niu, W., Cao, X. & Liu, Z. Effect of standard heat treatment on the microstructure and mechanical properties of selective laser melting manufactured Inconel 718 superalloy. *Materials Science and Engineering: A* **644**, 32–40 (2015).
28. Kirka, M. M., Medina, F., Dehoff, R. & Okello, A. Mechanical behavior of post-processed Inconel 718 manufactured through the electron beam melting process. *Materials Science and Engineering: A* **680**, 338–346 (2017).
29. Lu, Y. *et al.* Investigation on the microstructure, mechanical property and corrosion behavior of the selective laser melted CoCrW alloy for dental application. *Materials Science and Engineering: C* **49**, 517–525 (2015).
30. Fujieda, T. *et al.* CoCrFeNiTi-based high-entropy alloy with superior tensile strength and corrosion resistance achieved by a combination of additive manufacturing using selective electron beam melting and solution treatment. *Materials Letters* **189**, 148–151 (2017).
31. Zhou, K. *et al.* A precipitation-strengthened high-entropy alloy for additive manufacturing. *Additive Manufacturing* **35**, 101410 (2020).
32. Lin, W.-C. *et al.* Microstructure and tensile property of a precipitation strengthened high entropy alloy processed by selective laser melting and post heat treatment. *Additive Manufacturing* **36**, 101601 (2020).
33. Lu, Y. *et al.* Controllable Additive Manufacturing of Gradient Bulk Metallic Glass Composite with High Strength and Tensile Ductility. *Acta Materialia* 116632 (2021).
34. Zhang, Y., Chen, W., McDowell, D. L., Wang, Y. M. & Zhu, T. Lattice strains and diffraction elastic constants of cubic polycrystals. *Journal of the Mechanics and Physics of Solids* **138**, 103899 (2020).
35. Carroll, B. E., Palmer, T. A. & Beese, A. M. Anisotropic tensile behavior of Ti–6Al–4V components fabricated with directed energy deposition additive manufacturing. *Acta Materialia* **87**, 309–320 (2015).
36. Rodrigues, T. A. *et al.* Wire and arc additive manufacturing of HSLA steel: Effect of thermal cycles on microstructure and mechanical properties. *Additive Manufacturing* **27**, 440–450 (2019).